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ABSTRACTS

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CERAMICS

3-44



ELASTIC MODULUS AND INTERNAL FRICTION OF $\text{Si}_3\text{N}_4/\text{BN}$ FIBROUS MONOLITHIC CERAMICS

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Temperature and amplitude dependencies of Young's modulus and internal friction have been studied in Si_3N_4 and BN ceramics and composite materials $\text{Si}_3\text{N}_4/\text{BN}$ which were fabricated by hot pressing of the Si_3N_4 filaments clad with BN. The fibres were aligned along [0°], across [90°] or along/across [0°/90°] the axis of the rod samples which were used in the experiments.

The measurements of the Young's modulus E were made during the thermocycling in the 20-400°C temperature range. It was found that the high elastic modulus specimens of Si_3N_4 ($E=311$ GPa at 20°C) demonstrate a good thermal stability: the temperature dependencies $E(T)$ coincide each other at heating and cooling. The low modulus ceramics BN ($E=36$ GPa) shows a large hysteresis in $E(T)$ curves. That means that there exist some changes in the defect structure of the material which arise due to thermoelastic (internal) stresses. Qualitatively the same Young's modulus hysteretic behaviour is found for composite materials in spite of the high modulus filaments Si_3N_4 which make the Young's modulus higher up to 128 GPa for [90°] sample, up to 235 GPa for [0°/90°] and up to 295 GPa for [0°] samples. Note that the temperature hysteresis for the elastic modulus is observable only in the case of a microplastic deformation under the internal stresses.

The microplastic strain for the all materials under study is confirmed by the amplitude dependencies of internal friction and Young's modulus: the dependencies are similar to the curves measured for plastic metals and alloys.

The available experimental data are discussed in the framework of a model in which the temperature dependencies of elastic modulus and their peculiarities are explained by both non-linear oscillations of atoms in the lattice that are influenced by the internal stresses and microplastic deformations.

The work is supported by Russian Foundation of Basic Researches (grant 00-01-00482), NATO (grant PST.GLG.977016) and Energy Department of USA (contract W-31-109-Eng-38).



POROSITY AND MECHANICAL PROPERTIES OF CEMENT

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Cement (cement stone - CS) which is one of the most widely used building materials can be regarded as one of the types of ceramic. The SC structure is characterized by the presence, in addition to crystalline regions, of gel regions consisting of fine strongly distorted crystals. The properties of SC are determined in many respects by its porosity which can be as high as 20-30 %, the pore sizes ranging from several nanometers to fractions of a millimeter. The mechanical properties, cold resistance, and moisture permeability depends not only on the integral porosity of the material but also on concentration of pores of a particular size ranges. Therefore, the goal of this work was to obtain a clear picture of the pore size distribution for the entire range of sizes and also to reveal the relation between the porosity and the material properties, and, first of all, mechanical properties.

To solve the first part of the problem, a set of techniques – proton magnetic resonance, small-angle X-ray scattering, mercury measurements, optical and scanning electron microscopy – was used to determine parameters of pores of different sizes. In the investigations, integral and differential pore size distribution for the pores ranging in size from 2 nm to \approx 1 mm were obtained. It is essential that each technique was used for the pore size range where it proved to be most efficient. The obtained results were controlled by integral methods.

Such mechanical properties of SC as compression and tensile strength at different temperatures, effective modulus of elasticity, and microhardness of the crystalline and gel components were determined. A high hydrostatic pressure of up to 1 GPa was used as a factor allowing the SC porosity to be varied without changing other structural parameters.

It was shown that pressure substantially reduces the SC porosity; and the larger the pore, the higher the degree of its healing. By changing the pressure, we obtained information on the mechanical properties mentioned above depend on porosity. The remaining parameters of the structure (crystal size, degree of crystallinity, and so on) were found to be nearly unaltered by pressure. Possible reasons for a complicated behavior of a number of mechanical properties under low pressures (<0.1 GPa) are discussed, and thermal activation analysis of temperature dependences of the SC strength is given.



BENDING STRENGTH OF A PARTIALLY STABILIZED ZIRCONIA CERAMICS

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In previous works an exceptional for hydrostatically compressed crystals effect had been revealed in single and polycrystals of partially stabilized zirconia. This effect consists in decreasing in fracture toughness within pressure range up to 0.4 GPa. The effect was explained in terms of mechanisms of transformation toughening. A present work is devoted to studying strength characteristics of zirconia ceramics under mechanical bending testing. Ceramic specimens sintered from nanocrystalline zirconia powder stabilized with 3 mol.% yttria have contained almost 100 % of tetragonal phase.

Fracture toughness was measured by way of micro-indentation on the compressed or extended sides of the specimens under four-point bending. One of diamond indentation diagonals was oriented in parallel to the specimen's bending axis. Bending stress values were varied within ± 120 MPa. Fracture toughness factor K_{Ic} was calculated using conventional formulae separately for each crack propagation direction.

The dependences $K_{Ic}^n(\sigma)$ for cracks growing along the specimen's bending axis and $K_{Ic}^\perp(\sigma)$ for cracks growing across this axis have been obtained. Non-monotonic behavior was established experimentally.

As an elastic theory predicts, the longitudinal tension causes compressive forces in orthogonal directions. Taking into consideration this commonly known fact, one can suppose that the observed experimental strengthening, i.e. increase in fracture toughness factors K_{Ic}^\perp and K_{Ic}^n , in the ranges of negative values of applied bending stress σ (extended side of a specimen) or respective positive values of σ (compressed side of the specimen) is probably due to mechanical action of applied stress components impeding crack propagation (and/or nucleation).

The same mechanism can be used to explain the large drop in K_{Ic} within ranges of high negative σ (K_{Ic}^n) and high positive σ (K_{Ic}^\perp): in this case the tension stresses would be applied to the crack surfaces promoting the crack propagation.

Originally discovered peaks in curves $K_{Ic}^n(\sigma)$ at $\sigma = -50$ MPa and $K_{Ic}^\perp(\sigma)$ at $\sigma = 75$ MPa are of special interest. Since the sign of stress components determining mechanical action is favourable for crack propagation, the $K_{Ic}(\sigma)$ peaks observed are probably due to solely intensification of the tetragonal-to-monoclinic transformation. The strengthening effect on extended side of the specimen established in present work is well consistent with studies in hydrostatic pressure effects on fracture toughness. Applied pressure hinders the tetragonal-to-monoclinic transformation lowering thereby the fracture toughness factor. Consequently, 'facilitation' of the tetragonal-to-monoclinic transformation would lead to increase in the fracture



toughness factors in the case of prevailing nature of transformation toughening mechanism.

A number of experiments are evidential in favour of above suggestion. In particular, an increased content of monoclinic phase at extended side of specimen made from partially stabilized zirconia was found in work [1] using X-ray techniques while optical studies had revealed imperfections on the bent polished surface (it must be noticed that the monoclinic phase has specific volume about 5 % larger than tetragonal one).

Another direct method that can prove the growth in monoclinic phase content in mechanically extended areas is the micro Raman spectroscopy [2]. It is especially interesting that the phase composition was investigated in [2] under diamond indentation, that is, under loading mode similar to one used in present work, strongly convincing in validity of the suggestion made above.

In conclusion, microindentation experiments carried out in the present work on complex stressed specimens of partially stabilized zirconia have shown that a discovered strength anisotropy is caused by simultaneous action of two mechanisms, namely, mechanical action of applied stress on crack surfaces and intensification of tetragonal-to-monoclinic phase transformation in tension stress field.

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Y-PSZ FRACTURE TOUGHNESS ALTERATION DUE TO MECHANICAL PRE-LOADING

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It is known [1] that the high fracture toughness in ceramics based on ZrO_2 solid solutions (partially stabilized zirconia, PSZ) is caused by tetragonal-to-monoclinic martensitic transformation. Eigen value of fracture toughness in PSZ ceramics in absence of the martensitic transformation is about $1.1 \text{ MPa} \cdot \text{m}^{1/2}$ [2]; consequently, the most part of measured toughness magnitude is connected with this transformation. During mechanical testings, the martensitic transformation is initiated by a stress field of propagating crack, the stresses reaching values close to the material's theoretical strength. However, so high stresses are not necessary for initiation of themartensitic transformation.

The phase transformation initiating stress can be lowered for example as follows: a notched specimen is loaded up to K_I value less than K_{Ic} . Considering that the stresses reach the material's theoretical strength at load corresponding to K_{Ic} , one can lower the maximum arising stress by way of loading up to chosen K_I value. Following exposure at some K_I level the specimen is unloaded and tested to determine K_{Ic} . The similar experiments had been carried out in [3] in two modifications: the mechanical pre-loading was conducted at either constant exposure and various K_I levels or various exposures and constant K_I level. A monotone increase of K_{Ic} at rising pre-loading parameters has been revealed in both cases.

A studies in effect of the mechanical pre-loading on the K_{Ic} values are described in the present work. A subject of the work was Y-PSZ specimens manufactured from the powder YZ 01 of Saint Gobain Co. Measured density of the ceramics was 6.00 g/ccm that corresponds to relative density about 98.5 %. As for structure, the material can be characterized as $ZrO_2 + 5 \text{ wt.\% } Y_2O_3$ solid solution containing 1 wt.% ($MgO + CaO + TiO_2$). Mechanical pre-loading was imposed during 24 hours at K_I levels from zero (reference specimens tested without any pre-loading) to $9 \text{ MPa} \cdot \text{m}^{1/2}$ (fracture toughness level of the reference specimens). An effect of the mechanical pre-loading level K_I on a fracture toughness is shown in Fig.1.

It can be seen from the figure that in this case a dependence of fracture toughness on K_I is characterized by the complicated multi-stage curve. This complicated behavior can be due to increase in the transformation extent during mechanical pre-loading that can be possibly accompanied by a crack growth from the notch tip.



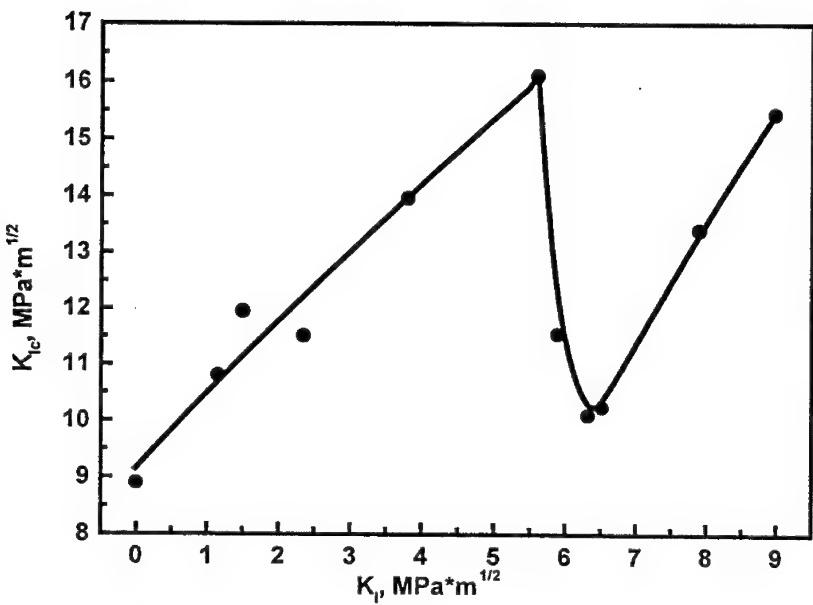


Fig.1. Dependence of K_{Ic} on pre-loading level K_I for PSZ ceramics at pre-loading exposure 24 hours

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FRACTURE TOUGHNESS OF CERAMICS AND CERAMIC COMPOSITES AT DIFFERENT TEMPERATURES

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Fracture toughness of oxide (alumina, zirconia, perovskites) and non-oxide (silicon nitride, silicon nitride with silicon carbide, silicon carbide with zirconium diborides and others) ceramics was investigated using Single Edge V-Notch Beam (SEVNB) method in a wide temperature range. The procedures for hand and machine polishing—out of the V-notch and the tests in three- and four-point flexures were created.

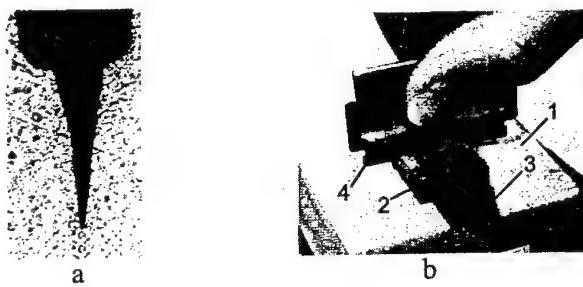


Figure 1. V-notch with a 5 μm root radius on a Y-PSZ specimen (a) and its manual polishing out (b): ceramic plate (1), additional specimens (2), specimen package (3), and razor blade.

The determination of load-deflection diagrams of V-notched specimens was used for a deeper insight into the deformation behavior of the ceramics studied at room temperature and at 1300–1400 °C. The data obtained by SENB and SEPB methods (Table) and micro-Raman spectroscopy (Fig. 2) were used in the analysis of the results. For instance, as was revealed, in the case of zirconia, if a V-notch root radius exceeds 5–7 mm, the results obtained coincide with those obtained by the SEPB method; polishing—out of the V-notch induces no tetragonal-monoclinic phase transformation in those materials; a slow crack growth was observed in the silicon carbide matrix composites, and so on.

Table. Fracture toughness obtained by SEVNB and SENB methods



Material	Fracture toughness, Mpa m ^{1/2}		Brittleness measure χ
	SEVNB	SENB*	
Soda Limit Glass	0.66±0.07	1.18±0.09	1
SSiC	2.61±0.18	4.42±0.23	1
Y-PSZ	5.14±0.29	9.54±0.47	1
Si ₃ N ₄	5.17±0.06	9.12±0.29	1
Si ₃ N ₄ +30%SiC+3%MgO	2.27±0.14	2.49±0.16	0.88
Mg-PSZ	9.44±0.12	10.2±0.27	0.41
<u>La_{0.2}Ca_{0.8}CoO₃</u>	2.2	2.2	0.25

* The width of notch is 0.2 – 0.3 mm.

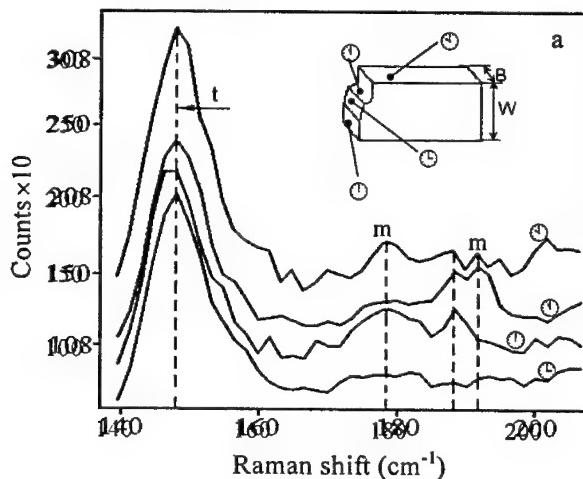


Figure 2. Raman spectra of a Y-PSZ specimen fractured by SEVNB method

It was also shown that the SEVNB method d it is very efficient in the field of material science.

THE ANALYSIS OF PROBABILITY OF FRACTURING OF CERAMIC INSERTIONS IN COMPOSITE CONSTRUCTIONS

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The problem of designing of the most rational forms of milled ceramic insertions widely used in a modern orthopedic odontology for restoring of occlusal surfaces of teeth, was already considered from the point of view of an estimation of strain intensity arising in a loaded constructions [1].

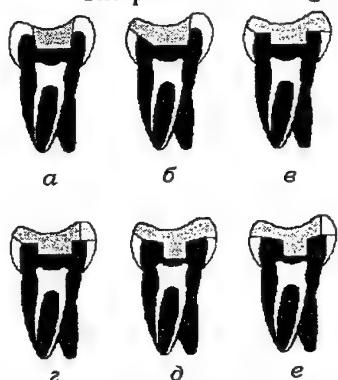


Fig.1. Forms of insertions: a - Inlay, b - Onlay, c - Overlay 1, d - Overlay 2, e - Pinlay 1, f - Pinlay 2

till two modifications of tabs Overlay executed with the SPLEN-K (INZOMA) system, oriented on calculation unhomogeneous multiply connected constructs. A finite element method is the mathematical basis of a system. The theory of fracturing Shleicher-Nandy is applied for an estimation of strength of a biomechanical construction, a parabola and ellipse were used as limiting curves.

For each form of insertions two schemes of a loading applied: a distributed loading simulating chew-

Let's consider one more approach to a solution of this task with usage of Shleicher-Nandy fracture criterion, permitting to construct more precise model of fracturing which is taking into account strength indexes of materials a constructions on squeezing and on a rupture.

For holding the comparative analysis the models of main aspects of ceramic insertions used in a modern odontology applied: Inlay, Onlay, and also and Pinlay (fig. 1). The calculations were

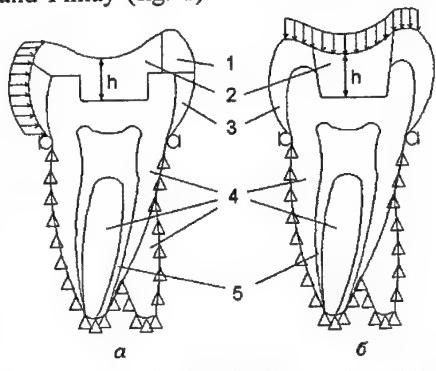


Fig.2. The calculated scheme of model of a tooth with a ceramic insertion: 1 - seal, 2 - ceramic tab, 3 - enamel, 4 - dentine, 5 - pulp



ing pressure, operating on a surface of a tooth in the field of chewing side and a unfavorable side loading from the buccal side (fig. 2). It first case an angle of an application of a load vary from 15 up to 165 degree, in second - from 105 up to 255 degrees. For all types of insertions the calculations of probability of fracturing of model of a tooth were carried out at the different schemes of a loading and angles of an application of a load.

The most unfavorable angle of an application of a load on a chewing surface makes 165 degrees. At a distributed load from the buccal side the greatest danger is represented by angles 105 and 255 degrees.

Fig.3. Different positions of a ceramic insertion and tooth dentine junction

For all forms of insertions the analysis of association of probability of fracturing of model of a tooth from height of a insertion h (fig. 2) was carried out. Height of a insertion h changed from 2 up to 4,5 mm with a pitch of 0,5 mm. The calculations show, as for a load on chewing side and for a load from the buccal side the greatest safety factor construct has at h component 3-3,5 cms for all forms of insertions. The considerable influence to a structural strength can also render a site of the junction between dentinomas of a tooth and ceramic insertion (fig. 3). For models of a teeth with insertions such as Overlay 1, Overlay 2, Pinlay 1, Pinlay 2 the analysis of association of probability of fracturing of model of a tooth from a site of the junction between a ceramic insertion and dentinomas was carried out at an operation of a functional load originating in the process chewing of nutrition. The calculations have shown, that most favourable is the position 3.

Thus, the calculations as a whole confirm the datas obtained as a result of a tension, carried out before the analysis, of model of a tooth with a ceramic insertion, though the scatter of values in limits 20 % is on occasion watched. The values obtained at application as limiting curve criterion Shleifer-Nandy of a parabola and the ellipse, differ no more than on 1 %.

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**MICRO-STRUCTURE AND PHYSICAL-MECHANICAL PROPERTIES OF
 Al_2O_3 -20%TiN COMPOSITES PRODUCED BY IMPULSE LOADING OF
 NANO-POWDERS**

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Composites Al_2O_3 -20%TiN are quite «attractive» materials as their components are one of the most widely spread compositions, methods of their ultra-disperse powders production are available enough and they have certain prospects in their application.

The purpose of this work was investigation of laws of structure-forming and destruction of experimental samples Al_2O_3 , TiN and also composites Al_2O_3 -20%TiN produced in conditions of impulse loading of nano-powders.

Mixing and homogenization of the mentioned composition were carried out in ethyl alcohol during 30 minutes. Drying was carried out in vacuum drying cabinet during 2 hours at 200 °C. Then experimental samples were preliminary compacted by conventional static method at pressure of 0.4 GPa without any binders. The produced blanks were compacted with shattering blasting agents in air according to plain loading scheme within a compaction interval of 3-7 GPa.

We have investigated phase composition, micro-structure and dispersivity of initial powders Al_2O_3 and TiN of different origin and also their influence on Al_2O_3 -20%TiN composite properties, which was produced in static and dynamic conditions. To determine quality and quantity phase composition of composites we were using «X-RAY», «AIST» and «PDWIN» programs of x-ray phase analysis automatic control.

It was found out that composites based on the most disperse and homogeneous electric-blasting and hydrolysis Al_2O_3 powders have higher relative density after conventional static compaction ($\rho_{\text{rel}} \sim 50\%$) and also after dynamic one ($\rho_{\text{rel}} \sim 84\%$). The lowest relative density had compacts of composites from plasma-chemical synthesis powders (Al_2O_3 and Ti), correspondingly, 36% - for conventional static compaction and 80% - for dynamic one. Composites were always compacted worse (to lower relative density) than initial components which was due to different shape of particles (spherical and polygonal) that make a composition. Density and strength of static compacts is very low (relative density 36-49%). There was mentioned a little change of phase composition of the components of the aluminium oxide after dynamic treatment. In a composite with hydrolysis powder there was observed 6% reduction of α -phase content and the same increase of θ -phase.



There was evaluated influence of different sintering regimes in air and vacuum within the temperature range of 1300-1600 °C on density of nano-composites produced by static and dynamic methods. It was shown that as sintering temperature grew density of composites compacted by explosion grew as well from 82-87% to 90-97%.

Micro-structures of cracks of composites samples compacted with explosion method and sintered in different regimes were studied by methods of electron microscopy. It was shown that in samples sintered at low temperatures (~ 1300 °C) cracks can be observed which, probably, remained from initial compacts. At sintering temperatures of 1500-1600 °C healing of such micro-cracks took place. Minimal size of composite grains was 100-150 nm. Increasing of temperatures lead to increase of a grain size to 1 mkm and higher.

Micro-hardness of samples produced by dynamic loading, sintered in vacuum at 1600 °C was maximal for composites based on electric-blasting and hydrolysis powders and was 25-30 GPa. Stickiness destruction coefficient K_{IC} was maximal also for samples produced in he same conditions and was $4.7\text{-}5.2 \text{ MPa}\cdot\text{m}^{1/2}$. Thus, H_V and K_{IC} for samples produced by dynamic method were 1.2-1.4 times higher than for standard ones, which, probably, is connected with higher density of a material produced by high-energy methods.

With the help of automatic program for calculation fine «GOR» structure there was carried out calculation of effective size of crystallites at the absence of micro-deformations influence and choosing of model of density length distribution of a number of cells and also calculation of distribution of micro-deformations at the absence of dispersivity influence and choosing of a model of micro-deformations distribution along the length in a crystal lattice. A joint solution of these two problems in approaching to isotropy was carried out.



BIPHASIC CERAMICS BASED ON CALCIUM PHOSPHATES

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At the present, materials based on calcium phosphates find wide application in medicine - ceramics of hydroxyapatite (HA) $\text{Ca}_{10}(\text{PO}_4)_6(\text{OH})_2$ and β -tricalcium phosphate $\text{Ca}_3(\text{PO}_4)_2$ (β -TCP). These phosphates are the basic inorganic component of hard tissue of a human organism. They show perfect biological compatibility and are well integrated with the bone tissue. Recently, special attention is given to production of biphasic ceramics of HA/ β -TCP with the ratio of 3/2, that is close to the ratio of these phases in bone tissue of a human. Such a requirement is caused by resone that stoichiometric HA, as an implant material, is rather slowly dissolved by the body fluid, and β -TCP - rather fast. The application of biphasic ceramics allows to find an optimum correlation between solubility of the ceramic forms of calcium phosphates and ingrowth of new bone tissue during implantation that makes rather perspective for practical use.

In the present work, a way for production of biphasic ceramics, in which the contents HA can be changed from 100 % to zero is proposed. The idea is to use HA as an initial powder and during annealing by "tear off" of OH⁻ group transfer its certain part in molecules β -TCP. As substance which is carrying out "tear off" of OH⁻ groups from a molecule HA ammonium sulphate $(\text{NH}_4)_2\text{SO}_4$ is chosen. The treatment of the HA powder by a solution of ammonium sulphate of different concentration should result in formation of a powder suitable for reception of biphasic ceramics HA/ β -TCP. The ceramics β -TCP was received in a similar way [1]. But for its preparation a mix of powders HA and ammonium sulphate was taken. The final product was a result of reaction in solid phase . For reception of 100 % β -TCP such an approach, probably, is convenient, but when it is required to make ceramics with a given ratio of the specified phases, and these phases should be even distributed on all volume of a sample, so it is more preferable select an approach, which is offered in the given work.

The synthesis HA as suspension is carried out by known wet method of a joint precipitation [2]. At the final stage of maturing the deposit was decantated and divided into 7 parts. One of them was an initial, and the others six were processed by a water solution of ammonium sulphate. The processing consisted in keeping of a deposit during 10 days in a rinse solution of $(\text{NH}_4)_2\text{SO}_4$. The rinse liquid was replaced on fresh in every 3 days. Concentration of ammonium sulphate was changed from 0,5 to 3,0 weights. % with a step 0,5 weights. %. Drying of the received deposits was carried out in a drying case at the temperature 150° C. After



that a deposit was pounded in porcelain mortar and sifted through a sieve with the size of cell 100μ .

From the received in such away powders the tablets were pressed which were sintered then at the temperatures 800, 900, 1000, 1100, 1250°C . The contents of HA and β -TCP in sintered ceramics depended on concentration of ammonium sulphate and determined on X-ray researches. The results are shown in the table:

Concentr. $(\text{NH}_4)_2\text{SO}_4$, weight %	0	0,5	1,0	1,5	2,0	2,5	3,0
Concentr. HA, weight %	100	86	70	43	12	0	0
Concentr. β -TCP, weight %	0	14	30	57	88	100	100

Gravimetry, X-ray, infra-red and scanning electron microscopy research of powders and ceramics, received from them, are carried. The intervals of temperatures, in which there is a final transformation HA in β -TCP, are determined. The mechanisms resulting in such transformation are discussed. The mechanical tests are carried out.

Thus, by processing of deposit of HA by the solution of ammonium sulphate the powder is synthesized, at sintering of pressings from which at temperatures $1150-$

1250°C biphasic ceramics with the given ratio of phases is formed. The mechanical characteristics of such ceramics do not concede to those HA, i.e. the received biphasic ceramics quite satisfy the requirements, showed to implantates, which are intended for introduction in the loaded or partially loaded sites of a skeleton.

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POROUS HYDROXYAPATITE GRANULES WITH INCREASED MECHANICAL PROPERTIES

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Hydroxyapatite $\text{Ca}_{10}(\text{PO}_4)_6(\text{OH})_2$ (HA) - basic component of an inorganic component natural bone. If by calcination to get rid of an organic component, for example, in spongy bone, instead of him there is a network informed porous of two types: macroporous and microporous with the sizes in hundred and a few micron respectively.

HA recently has received a wide circulation in quality implant of a material in orthopedics and maxillary-facial of surgery. It is caused by that the marked chemical similarity provides HA a high degree of biocompatibility, bioactivity and osteoconductivity (germination of a fabric) in relation to natural bone.

In practice HA is used as ceramic dense and porous briquettes, granules and coated on metal substrate. Lack dense HA of ceramics is the duration of its transformation in alive bone, as in this process is realized only soluble-precipitative the mechanism and the mechanism osteoconductivity is not involved. For faster and effective healing bone of defects implant from HA of ceramics should be porous, i.e. have macro- and microporous of the mentioned above sizes. In case of application of such material because of increase of the effective area of interaction intact bone and implant the optimum conditions for the accelerated formation of a bone-ceramic complex as are created by course soluble-precipitative of reactions with formation new bone, and germination environmental bone in ceramics.

The basic lack porous HA of ceramics is its low mechanical durability in comparison with natural bone. As is known, the mechanical properties of ceramics can be improved by increase of its density. One of opportunities of increase of density of ceramics, including from HA, is use more dispersive of an initial powder.

For reception high-dispersive of a powder HA we a little modified a known method Jarcho. The difference was, that in a reactionary mix added superficial - active substance, which did not allow primary nanocrystals HA to be united in conglomerates. The powder, received by such method, had the sizes of particles less than $0,2 \mu$. IR-spectrometer and X-ray analyses have shown, that the given powder is HA and within the limits of sensitivity of the specified methods in it are not found out admixive radicals and phases.

In quality porophore used foamed polystyrene (FP). The size particles porophore was picked up so that after drying and anneal granules the size macroporous in it made $400\text{-}600 \mu$; it corresponds to the sizes macroporous in



spongy bone of the man. A powder HA carefully mixed with the selected fraction FP and, adding distillative water, received elastic paste. From small portions of this paste by running in between two parallel-sided by plates rotating in opposite directions, prepared preparations rounded of the form close to spherical, diameter of 6-8 mm. After drying at 100-120° C with within day these preparations subjected preliminary anneal (calcination) at 800° C with in pairs of water (in avoidance dehydrate HA and his transition in threecalciumphosphate), further temperature anneal lifted up to 1100-1150° C with and preparation anneal within 2 hours. During sintering the size of preparations decreased, and as a result received porous granules HA by a diameter of 4-6 mm.

The X-ray and IR-spectrometer researches have shown, that at anneal close 200° C with FP was decomposed to ammonia and benzol, which with increase of temperature evaporated. Above 440° C with in preparations FP already is not present, and the material granules after sintering represents HA without admixive of phases.

Made granules contained porous of two types: spherical macroporous by a diameter 350-600 μ which have arisen owing to burning out of balls FP, and microporous. Last were formed as a result of an output vapour of water at drying both products of decomposition and burning porophore in time anneal. They represented a network of the interconnected channels with the cross size in some microns. The rating microporous has given value within the limits of 25-35 % - compressive durability 35-40 MPa. Volume macroporosity could be lead up to 80 % and was set by quantity FP general porosity from 45 up to 50 % ($\pm 5\%$) compressive the durability is estimated 20-25 MPa that close to the appropriate characteristics spongy bone of the man.

Granules differed transparent porosity, that was shown by impregnation by liquids of various viscosity, simulative official preparations. Due to an acceptable combination of the marked functional characteristics received granules have good prospects for clinical use in quality fillers partially - are loaded and non-loaded bone of defects.



STRAIN RATE SENSITIVITY OF TZP-CERAMICS NANOHARDNESS MEASURED BY DYNAMIC NANOINDENTATION TECHNIQUE

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Tetragonal zirconium polycrystal (TZP) ceramics ZrO₂ is a unique material possessing both very high hardness and fracture toughness K_{IC} close to the steel ductility. However, the reasons for such a high durability of this promising material are not clear up to date. It is well known the materials with high magnitude of K_{IC} and low hardness (H) as far as those with high H and low K_{IC} subjected to the very fast wear. Therefore, some optimal combination of hardness and fracture toughness is necessary. At the same time we must deal with the dynamic hardness but not with the static one. Material behavior during dry friction, ball mills processing and abrasive wear can be modeled by means of the fast local straining under indenter at

high relative strain rate, $\dot{\epsilon} = \frac{1}{h} \frac{dh}{dt}$ ($10^3 \div 10^5 \text{ s}^{-1}$). High magnitude of $\dot{\epsilon}$ can be

achieved at very moderate linear rates of indenter penetration and low depth h of imprints. For example, it is necessary to have $dh/dt \approx 0,1 \div 1 \text{ sm/s}$ at $h \approx 0,1 \div 1 \mu\text{m}$ in order to achieve above-mentioned values of the strain rate. However, despite of the fact that time-dependent material properties were studied since 1970's, the only creep which takes place at $\dot{\epsilon} \sim 10^{-2} \div 10^{-5} \text{ s}^{-1}$ was understood in details. The milliseconds contact time range is still remaining almost unexplored.

For time-dependent hardness investigation the special apparatus was elaborated and built up in the nanoindentation laboratory at Tambov State University. The nanotester has high displacement (0,1 nm) and temporal (50 μs) resolution at wide load rate range (12,5 mN/s - 12,5 N/s).

It was determined that the dynamic hardness essentially depended on active loading duration ($10^{-2} \div 1 \text{ s}$) changing up to 70% for different TZP ceramics with the various grain size, composition and concentration of stabilizing impurities. At the same time, the stain rate sensitivity factor $m = d(\ln H)/d(\ln \dot{\epsilon})$ for all samples is strongly differed for initial and final stages of indentation process. For penetration depth $h_1 = 200 \text{ nm}$ the m factor magnitude was three times higher than

those for penetration depth $h_2 = 650 \text{ nm}$ (at $h_{\max} = 755 \text{ nm}$). Therefore, $\frac{m_1}{m_2} \approx \frac{h_1}{h_2}$.

Such essential difference for the stain rate sensitivity factor at various h brings an evidence of changing the dominating material plastic deformation mechanisms at different stages of indenter penetration. The curve $H = f(\dot{\epsilon})$ was analyzed in



double logarithmic coordinates. Two intercepts with an ordinate were found out upon extrapolation, $H_0' = 20,2$ GPa and $H_0'' = 8,0$ GPa. Such behavior of $H = f(\dot{\varepsilon})$ allow us to distinguish time-dependent and time-independent components of TZR ceramics hardness. Moreover, the sensitivity of the hardness to the strain rate can be directly determined from the P-h diagram shape. It is enough to compare sharpness of the P-h diagram peak near the P_{max} at different loading rates. The less duration of the active loading, the more equivalent curve radius.

Thus, the present study of some TZP ceramics is allow to extract the time-dependent hardness component that could be significant for understanding the behavior of these materials at short nanocontact interactions.



ANALYTICAL RESEARCHES OF COMPOSITE MATERIALS WITH BRITTLE MATRIX AT STATIC AND CYCLICAL LOADS

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In last 30 years are conducted researches on creation of details and units of high-temperature turbines from structural ceramics. However integrally proper to this class of materials the brittleness constrains broad application them in engineering. One of the conventional ways of increase of strain crack toughness of ceramics is the creation of ceramic composite materials. This class of materials is considered perspective for application in details both units of air and space engineering working in conditions of high temperatures. In spite of the fact that the intensive researches are conducted in the field of ceramic composite materials, the analytical and experimental results are represented poor for construction of stringent mathematical models of their deformation. One of observable features of ceramic composite materials is that the limiting deformations of a material of a matrix appear below, than material of filling materials. Besides a porosity of a material of a matrix, residual stresses in an off-the shelf article, and also other factors can determine behaviour of parts from composite materials at their cyclical and static loading. In this connection the analysis of influence of a porosity of monolithic ceramics on a its compliance and strength was carried out. The results of these researches are used for construction of model of deformation of a ceramic composite material with a brittle matrix.

Influence of a porosity on a compliance and stresses of specimens from monolithic carbide of silicon (SiC) was explored originally. At this the prismatic sample by the size 25x5x2.5 with a random distribution 76 spherical pores of diameter 200 microns was considered. Is shown, that at three point bending the compliance of a sample has increased on 12 percents, whereas the maximum values of stresses have remained practically at a former level. Besides is established, that the greatest localized stresses are localized in neighbourhood of surface pores located in the field of maximum tensile stresses. At last, by analytical researches is confirmed, that for current value of principal stresses there is such size of a pore, when in one of neighbourhood the maximum values of stresses concentrate. At some increase of the size of such pore the level of maximum values of stresses is a little reduced. This circumstance causes a capability of slow growth of a crack up to some critical level. All said has allowed to offer version of mathematical model of deformation of composite materials with a brittle matrix grounded on application of methods of a mechanics of elastic - porous mediums and a mechanics of destruction. First of all, on the basis of the theory of elastic - porous mediums it is supposed, that the elastic modulus E^2 of a material of a matrix decreases with growth of the fraction contents of pores. At static and cyclic loads the increase of the volumetric contents



of pores is possible. It is supposed, that the change of a porosity of a material of a matrix is a degree function of a factor of intensity of stress. Besides known in a mechanics of failures a ratio linking a factor of intensity of stress with the operational stress and the size of a pore is used. At an estimation of a level of the operational stress the residual stresses stipulated by used technique of manufacturing can be taken into account. The relation of a porosity to time is established after an integration of a kinetic equation of change of a porosity. Thus, the actual relation of elastic modulus E^2 from time and elastic - viscous characteristics of a material of a matrix is analytically determined. On the basis of the widespread approaches the estimation of the effective elastic characteristics of unidirectional - reinforced ceramic composite materials is carried out in depending on the fraction contents and properties of their substituents (filling material, matrix). As the elastic modulus E^2 of a material of a matrix depends on a level of the operational stress, the effective characteristics of a composite material also depend on a level of the operational stresses. This circumstance determines stipulated by slow growth of a porosity non-linear area of the analytically constructed single-axis curves of deformation of composite materials with a brittle matrix. The possible consequent failures on a interface of substituents of composite materials or reinforcing filling are considered. The qualitative and quantitative comparisons with existing experimental data are carried out.



TECHNOLOGY FACTORS ON FORMATION PROPERTIES of POWDER STUFFS

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One of perspective directions of creation of new technologies of mechanical engineering effecting is obtaining bars for structural parts from powdered materials. At designing of master schedules of manufacturing of bars there is a series of tasks, which one are connected to obtaining of indispensable operating characteristics of parts. By the relevant task, which one is decided at an intrusion of technology, there is obtaining the given mechanical characteristics of a stuff. For obtaining a stuff with satisfactory strength, wear resistance and other service properties it is necessary to give him an inner pattern, which one would respond a task of obtaining of the conforming level of the mechanical characteristics. It is possible to reach through optimization of parameters of a master schedule. However creation of optimum technology is impossible without an estimation of influencing of technological parameters on a qualitative condition of a powder stuff.

For calculation of the mechanical characteristics of porous materials stuffs offered relations [1], which one are grounded on the count of influencing of basic performance of a powder stuff by which one there is a porosity. The equations allow for a level of reduction of strength of a stuff in a consequent of the obtained porosity accordingly by relation of a porosity to the characteristic of strength of a stuff in a compact condition.

The research of processes of plastic deforming of powder stuffs allows to determine their actual mechanical characteristics, to forecast modes of master schedules for obtaining stuffs and items with given properties. The definition of actual properties of porous materials stuffs and analysis of their behavior at deforming is a main problem of powder metallurgy.

At research of mechanical properties of powder stuffs it is marked, that near to influencing of a porosity on strength essentially influence as well technological parameters. Appreciable value on process of obtaining of indispensable strength of a powder stuff have such technological parameters, as a method of obtaining and grain composition of output dusts, methods of their consolidation, temperature both duration of a sintering treatment and other conditions of manufacturing. The variation of technological parameters allows considerably to influence an inner pattern of a stuff. The random nature of influencing of grain composition on processes of consolidation, on formation of mechanical properties of a powder stuff puts the indispensable sanction of a task of definition and count of influencing of technological parameters. Allowing this alignment of definition of mechanical



properties under influencing of a porosity and technological parameters should be determined by a function of a kind

$$\cdot f(F, T, Q, P)$$

where F - parameter, which one depends on distribution of grain composition;

T - parameter of reliability of formation of contacts, which one depends on sintering temperature and thermal exposure;

Q - parameter of a porosity;

P - parameter, which one allows for sequence and kind of technology of obtaining of a stuff.

Frame and the properties of powder stuffs are determined by combination of many technology factors. From the point of view of maintenance of reliability of obtaining of given properties of a stuff the considerable influencing has grain composition of output dusts. For definition of a role of grain composition a series of experiments with dusts iron powder IDKP3 was conducted. For researches the fractions of 0,063 mm were excreted, and 0,2 mm, that were screened through the screen. Pressing characteristics determined by formation cylindrical is model with a ratio of length to diameter 1,5. Samples were produced from a blend of a dust with fragments of one and two fractions. Samples were produced 4 groups. Their formation was conducted in the steel press form by a method of bilateral pressing. The blend before a filling was weighed to receive samples of an identical porosity. After a sintering samples were deformed on compression. By results of researches, the function is offered, which one allows for influencing grain composition on a porosity of a stuff by the way

$$Q = 1 - \exp(-APd)\rho_{D_{max}}$$

Where - $\rho_{D_{max}}$ - relative consistency of a stuff from a dust of a large fraction;

Pd - percentage of a large fraction;

A - factor. Value A=0,0196

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**APPLICATION AMPLITUDE-DEPENDENT INTERNAL
FRICTION TO ANALYSIS OF PLASTIC BEHAVIOUR OF ZIRCONIA.**

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Investigations into non-elastic effects provide data of importance on the defect structure of solids. The non-elasticity as a property of a real solid body resulting in a non-simple stress-strain relation is manifested itself as two closely related effects: the internal energy dissipation and the elasticity modulus defect (EMD). An alternating external stress applied to a solid body may cause different mechanisms to be in action, but their contributions into the total damping are not equal and depend on the loading conditions. Amplitude-dependent internal friction (ADIF) may provide information on the type of the non-elastic phenomena mechanism, when the IF and EMD are studied in parallel [1,2].

Before [1-3], we have studied the dislocation IF, EMD in Zn and Mo single crystals. The present work is dedicat to investigations into non-elastic effects in zirconia. The ZrO_2+3 mol.% Y_2O_3 powder obtained by a method of coprecipitation from a solution of salts $ZrO(NO_3)_2$ and $Y(NO_3)_3$ by a water solution of ammonia. The mass part of impurity was: The $3,5 \times 3,5 \times 40$ mm specimens for acoustical testing were recived by sintering (1450^0C , 2h) of the green compacts hydrostatically pressed from dry powder with pressure from 100 up to 1000 MPa [4].

The IF - coefficient δ and the elasticity modulus defect $\Delta M/M$ in zirconia crystals were measured simultaneously using the doubled compound vibrator at the longitudinal vibration frequency about 91 kHz. The δ measurement accuracy was about 6%, that of $\Delta M/M$, about 2×10^{-3} %. The measurements were made in temperature range from 293 to 473K at fixed T values being set and maintained to within 0,1K.

The amplitude-dependent decrement for increasing the strain amplitude in the zirconia is in contradiction to the Granato-Lukke theory. Therefore in present work a phenomenological theory of the amplitude-dependent IF in the range of stress critical amplitudes is proposed. The theory is based on the representation about possible description of the inelastic behaviour of materials by means of the mechanical hysteresis loop [5]. Expressions for the amplitude dependences of the oscillation decrement and elasticity modulus defect have been obtained for differend kinds of the hysteresis loops.

If the hysteresis loop is a parallelogram, amplitude-dependent internal friction in the range of critical stress can easily be obtained that:



$$\delta(\sigma) \approx 4 \frac{M}{M_d} \frac{\sigma - \sigma_{cr}}{\sigma_{cr}}$$

amplitude-dependent elasticity modulus defect:

$$\Delta M/M(\sigma) \approx \frac{4}{3} \frac{M}{\pi M_d} \left(\frac{\sigma - \sigma_{cr}}{\sigma_{cr}} \right)^{3/2}$$

For a double hysteresis loop amplitude-dependent internal friction:

$$\delta(\sigma) \approx \frac{M}{M_d} \frac{\sigma - \sigma_{cr}}{\sigma_{cr}}$$

amplitude-dependent elasticity modulus defect (ADEMD):

$$\Delta M/M(\sigma) \approx \frac{1}{2} \frac{M}{M_d} \frac{\sigma - \sigma_{cr}}{\sigma}$$

when: σ_{cr} - microyield stress, M_d - microyield hardening, M - elasticity modulus

Microyield stress σ_{cr} and microyield hardening M_d evaluated from the ADIF and ADEMD have been studied as functions of the temperature for ZrO_2 crystals.

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USE of $YCrO_3$ And ZrO_2 CERAMICS FOR RECEPTION of HEAT RESISTING COVERINGS ON CHROMIUM ALLOYS.

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On the basis of results of kinetics research of oxidation of chromium alloys and phase composition of scale formed on them, the system of a two-layer heat resisting covering for protection of chromium alloys from nitridation and oxidation including yttrium chromite ($YCrO_3$) and stabilized zirconium dioxide (92 % ZrO_2 + 8 % Y_2O_3) was determined. Yttrium chromite is used as heat resisting a antinitrogen layer, and dioxide zirconium - for giving the greater hardness and durability of coverings ($\sigma_{bend} = 500-600$ MPa for ZrO_2 (Y_2O_3), $\sigma_{bend} = 60-90$ MPa for $YCrO_3$) and increasing of resistance to erosion and corrosion.

It is known, that REM chromites and, in particular, yttrium chromite, at temperatures higher than 1273K at the presence of oxygen is exposed to incongruent evaporation chromium oxide, and modified zirconium dioxide with tetragonal structure at these temperatures is a rather stable material.

In the given work the results of researches of thermal stability of yttrium chromite alloyed of MgO , CaO , SrO , Al_2O_3 , ZrO_2 oxides are given.

The stability of a material was estimated by speed of weight loss Vq (g/m^2 h). Samples tested on air and in nitrogen, were weighed at once after cooling, and tested in Na_2SO_4 melting - after 5-6 multiple washings in boiling water and drying at 473-523 K. Phase structure of samples was analyzed on installation ДРОН-4 in K_α -Cr radiation x-ray tube БСВ-2.

As a result of the carried out researches it is established, that the highest increase of thermal stability of yttrium chromite is achieved by complex alloying 5 moth of % ZrO_2 (or 5 moth % MgO) and 10 moth % Al_2O_3 (table).

Under alloying of yttrium chromite by the same amount CaO or SrO , that is much higher their solubilities in yttrium oxide (1,5-3 moth %) is observed it destruction and increasing of weight loss (table). In its phase composition unlike the previous case at the same time with yttrium chromite there are phases β - $CaCr_2O_4$ ($SrCr_2O_4$), $CaCrO_4$ ($SrCrO_4$) and Cr_2O_3 . At alloying of yttrium chromite by higher amounts of zirconium, magnium, aluminium oxides, than marked above, the weight losses also grow, but in smaller degree, than under alloying of calzium or stronziun oxide. In phase composition it is found out at the same time with yttrium chromite cromium oxide and alloyed of elements oxides. It is necessary especially to note high thermal stability in the specified environments zirconium dioxide, which is higher on the order than optimum alloyed chromites (table, composition 2, 5)



Table.-Speed of weight loss (g/m² h) for different chromites in various environments.

№ n/n	Material	Air	Nitrogen	Na ₂ SO ₄ , 1023K, 25 h.
		1573K, 10 h.		
1	YCrO ₃	0,4	0,1	1,1
2	Y _{0,9} Zr _{0,1} Cr _{0,8} Al _{0,2} O ₃	0,07	0,03	0,15
3	Y _{0,9} Mg _{0,1} Cr _{0,8} Al _{0,2} O ₃	0,12	0,02	0,3
4	Y _{0,9} Ca _{0,1} Cr _{0,8} Al _{0,2} O ₃	2,8	1,5	6,2
5	ZrO ₂ + 8%Y ₂ O ₃	0,03	0,01	0,045

The alloying of yttrium chromite by zirconium, aluminium and magnesium oxides also rises its strength properties. So, for Y_{0,9}Zr_{0,1}Cr_{0,8}Al_{0,2}O₃ $\sigma_{bend}=80-130$ MPa and $\sigma_{bend}=60-90$ MPa are for YCrO₃. ZrO₂ + 8%Y₂O₃ ceramic was the best strength ($\sigma_{bend}=250-300$ MPa).

The investigated materials have a high level of heat resistance and can be used as a basis of heat resisting composite coverings. The given ceramic materials can apply for a target at magnetron dispersion, which rather precisely reproduces chemical composition of a material of a target.



ROLE OF BENDING MODE IN PROCESS OF FRAGMENTATION OF CRYSTALLINE SOLIDS

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At existing level of knowledge used conception of two plastic deformation modes (translation and rotary) it is difficult to explain number of structural changes which take place at intensive plastic deformation. In particular it is incomprehensible in how in one case fragments of polygonal form is forming, while in other case the stripe structure arising, and why overwhelming majority of boundaries of the fragmented structure does not connected to sliding planes.

New idea about mechanisms of mesolevel deformation is developed in a cycle of works the main list of which is placed in [1]. There it is proposed to introduce conception of new deformation mode, bending one, which forming the structure with smooth change of misorientation of regions of crystalline lattice. Such structures are described also in row of other works. Authors of this works are considering it as a rotary mode manifestation and interpreting this effect with assist of disclinations. But analysis of investigating is showing that dislocation structure evolution at great plastic deformations goes throughout stage of formation of crystal lattice mesoscopic local bends. It is proposed to consider the local bend as new mesoscopic defects of crystalline lattice and name it "displanation". In contradistinction to linear defects, dislocations and disclinations, the displanation is volumetric ones. It is established that in common case arised bends has dipole character, but in particular cases bends of one sign are possible. Bends on electron microscopy image are appeared in form of specific extinction contours. Components of bending-torsion tensor, which describe of observed lattice bend, can be determined by the electron microscopic methods. The dislocation model of displanation connecting the bend appearance with creating of two tilted dislocation rows of different sign is proposed [3].

At great deformation degrees the curvature of crystalline lattice can reach of tens grades per micron. In this case elastic stresses proportional to the density of dislocations in pill-ups are creating in zone of the bend. In certain moments these stresses became critical ones for a lattice. The exceeding of its leads to local violation of initial interatomic bonds on mesoscopic extent and instead of the smooth misorientation the discrete one arises. The analysis based on dislocation model of the local bend shows that in the case two types of boundaries are arised (Fig.): boundaries connected to formation of the dipole of partial disclinations (1 and 2) and boundaries arised as a result of shear which does not coincide with planes of crystallographic sliding (3). Last ones are like to boundaries, which are forming of martensitic crystals, and arise as result of the local shear of atoms along the line of



the maximum planes squeezing because of interatomic distance in relation to initial one leads to greater stresses than its increase. The relaxation of stresses going on inside dislocation charges as well: dislocations gather to center and forms, *i.e.* two disclination boundaries. It is necessary to note that real rotations in case are small enough, but a large misorientation degree is prepared by foregoing bending. Bends with one of components equal to nil leads to the formation of stripe structures. In case of planes distortion in three directions polygonal fragments are forming.

Thus, the local bend is a precursor of fragmentation and determines form and size of the forming fragments. Character and parameters of local bends and, correspondingly, future fragments depend, mainly, on crystallography of sliding, conditions of deformation (stresses state, velocity and deformation temperature) as well as initial structure and rigidness of the interatomic bond.

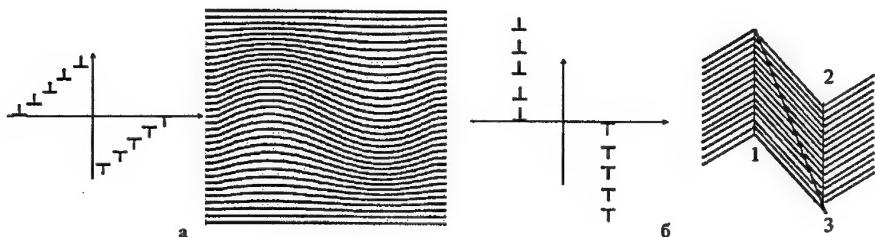


Fig. Scheme of the fragment boundaries:(a) bending and (b) rotary modes; (1, 2) dipole of disclination; (3) noncrystallographic shear.

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SUBJECT OF THE EFFECT OF INTERACTION BETWEEN COMPONENTS OF A CERAMIC COMPOSITE ON STRENGTH

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In practice all the qualitative conversions in technical equipment of a mankind are connected with inversions and a mastery of new materials. The XXI century will not be an exception and progress of modern technique will depend more and more on producing new materials. Such materials first of all are ceramic composites (CCs), based on a uniquely simple idea of combining various in strength components as it is observed in nature.

A successful solving of problems connected with CCs production requires a complex approach taking into account a variety of factors accompanying «the behavior» of CC components both in the process of producing materials and its exploitation.

The present work is dedicated to the development of such an approach using the possibilities of application of modern computer facilities and investigation technique. CC is considered as a functional substance, i.e. the substance (or a set of substances) in a definite phase state with an inner structure, which determine properties interval providing a realization of the given function. The work presents a topological CC description, including structural components in the form of hard ceramic grains, distributed in a «softer» ceramic matrix, and a metric one, including a description of quantitative parameters, determining a spatial structure of the substance, connected with technical peculiarities of the material production. In the technology of CC production the very process of making the material is ensured by a thermobaric factor acting in time. A role of this effect is determined differently for various compositions, however a general factor is the development of two competitive processes: 1) the increase of a binding strength between components (a development of interaction); 2) degradation of the initial components properties, particularly, recrystallization, origination of «harmful» layers of interaction (fragile ones), destrengthening of CC components. That is shown in the CC gamma:

silicon nitride - silicon carbide;
ultradispersive diamond - aluminium oxide - glass binder;
silicon carbide - titanium carbide;
aluminium oxide - glass component and some others.

Realization of the needed functionality in some cases comes to determination of optimum geometry of CC components. In this work there are found out theoretically and proved experimentally the dependencies of the listed



CCs on the bulk part (or thickness) of «harmful interlayers». Shown is a competence of the used conceptions when producing CCs with improved exploitation properties ensured by the presence of «weak» surfaces, realized in the process of CCs production. Therewith the process of production is considered as an independent functional control subsystem with the inner organization of the objects and interconnections of its own.

The result of the work is a solution of several practical problems, e.g. production of CCs and products thereof with the needed strength when operating on bending (cyclic loading). There is developed blade and porous abrasive tool from the listed components.

There are obtained valuable results enabling to make definite conclusions and to give practical recommendations as for making new materials.



INFLUENCE OF GRAIN BOUNDARIES COMPOSITION AND CONDITION ON THE PROPERTIES OF ALUMINA CERAMIC

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A microstructure - properties study of the hot-pressed pure alumina materials and alumina finegrained matrix composites containing 5 wgt % of nano-dimension nonoxide and oxide particles with different thermal expansion coefficients: SiC, Si₃N₄, SiO₂, ZrO₂ has been undertaken. The structure, hardness, fracture toughness, wet erosive wear and wear surface have been investigated.

It has been confirmed that composition and condition of the grain boundaries play an increasingly important role in nanomaterials, in correlation with their large volume fraction.

It has been demonstrated that aluminas hot-pressed at high temperatures (1600-1720°C) can contain pores inside grains due to grain boundary evolution and their fracture mode partially changes from intergranular to transgranular (Fig.1).

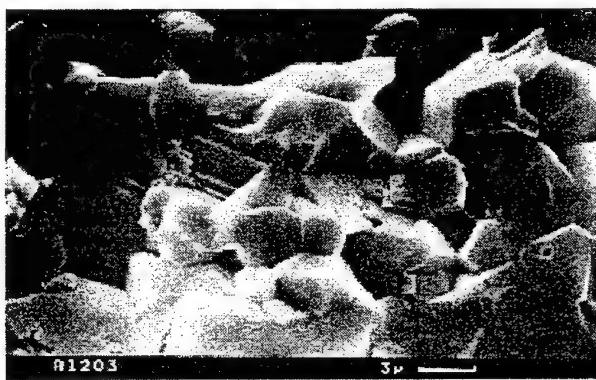


Fig.1 SEM of fracture surface of aluminas hot-pressed at high temperatures (1600-1720°C).

The fracture mode changed from predominantly intergranular for the pure alumina to predominantly transgranular in the nanocomposites. The difference



in the thermal expansion coefficients is not the main reason for this phenomenon. The best composite material containing 5 wt. % SiC particles exhibits transgranular fracture (Fig.2) and excellent wear resistance because the grain boundary composition is Si-Al-O-C.

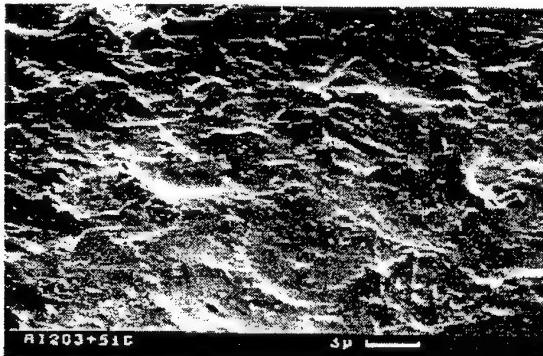


Fig.2. SEM of fracture surface of Al₂O₃-5wt.%SiC composite.

Composite materials containing 5 wt.% Si₃N₄ (Fig.3) or SiO₂ also exhibit predominantly transgranular fracture, but with a large proportion of intergranular cracking, so their wear resistance is lower. Perhaps this is because the composition of grain boundaries does not include carbon. The composite materials produced at the temperature 1700°C have the lowest wear rate when the dwell time of hot-pressing is short, leading to amorphous grain boundaries. By contrast, pure aluminas exhibit a decrease in the wear rate for low temperatures of hot pressing and long dwell times, resulting in ordered grain boundaries.

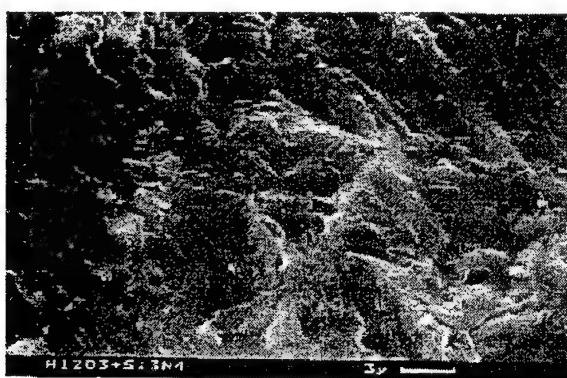


Fig.3 SEM of fracture surface of Al₂O₃-5wt.%Si₃N₄ composite.



STATISTICAL FEATURES OF MICRORACKING OF NON-ELASTIC CERAMICS

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The goal of the proposed work is to study the interrelation between structure, mechanical characteristics as well as fracture behaviour of non-elastic ceramics. It is essential for awareness of an optimal design and processing parameters during fabrication of the material.

For this purpose, the model of failure, which can be applied to n-phase brittle materials (in particular to ceramics) is considered here. The authors try to solve a physical problem of the description of non-elastic ceramic behavior as stochastic process of cracking of separate structural elements. The above mentioned model is applied for the description of mechanical behaviour of single-phase ceramics with various statistical distributions of the grain sizes. An effective continuum and a statistical description of the failure process are used.

Characteristics of the individual structural elements, the form of statistical distribution, and loading parameters are used as the initial data for calculating the volume fraction of the fractured structural elements. An arbitrary structural element is adopted for the effective continuum. Anticipation of the fractured structural elements volume and probability of the individual failed element are calculated with due consideration being given to the local failure criterion. The probability of the individually failed element is assumed to be proportional to the relative number of the fractured structural elements. The local failure criterion is selected to coincide with the energetic approach of crack nucleation [1].

The dependence of the normalized stress on the normalized strain for a single phase material is shown in Fig.1. There are four stages of process under condition of fixed stress (uniaxial tension). Really third and fourth stages are during very short time as result of fracture localization under tension. Hence, these stages are invisible in a practice under condition of fixed stress. Third stage is absent in this case. Second stage of scattered microcracking is up to the stage of catastrophic fracture. The present work is based on the refractory alumina ceramics with addition of zirconia (CZ) [2], fabricated from the conventional industrial raw materials. The maximum dimensions of its structure elements approximates to hundreds of micron. The statistic distribution of grain sizes can be easily approximated by the appropriate logarithmic normal distribution.



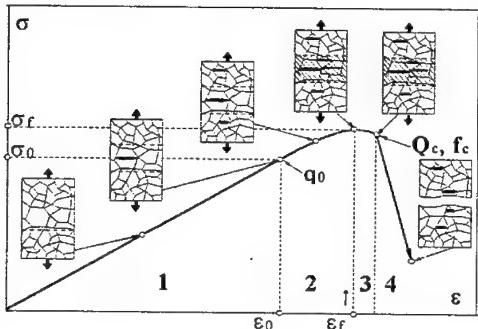


Fig.1. Strain diagram of the material with the local stochastic microcracking: 1 – loading without microcracking; 2 – a stable non-localized microcracking before stress maximum; 3 – a stable localized microcracking after stress maximum; 4 – unstable (catastrophic) fracture.

The true experimental stress-strain diagram and the calculated stress-strain diagram are presented in Fig.2. These diagrams are in good agreement up to stress maximum. Further differences are due to stress-induced phase transformation in zirconia particles. This process requires additional energy and it is additional dissipation of elastic strain energy. It results in slower microcracking and more essential strain reached in outer layer of specimen. This additional energy dissipation was not to take into account in our model calculation.

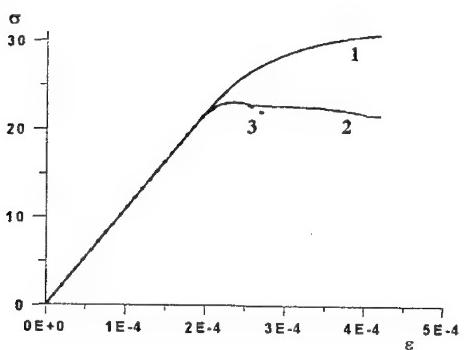


Fig.2. Stress - strain diagram of CZ ceramics: 1 – nominal; 2 – true; 3 – calculated.

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INVESTIGATION OF MECHANICAL PROPERTIES
AND DEFORMATION MECHANISM FOR WC-Co HARD ALLOYS
IN THE WIDE TEMPERATURE RANGE

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WC-Co alloys are used in the industry widely. But mechanical properties and deformation mechanism of these alloys at elevated and low temperatures are studied weakly. In the present work alloys with grain size from 0.25 to 2.3 μm and Co binder content 6, 10 and 15% were investigated. The structure of alloys was investigated by SEM with magnification up to $\times 40,000$ (fig.1). There were determined the average grain size d , mean free path in cobalt binder λ and contiguity of the carbide phase C. Hardness was investigated in the temperature range -196÷900°C at load 60 and 200 N (fig.2). Mechanical bending test were carried out in vacuum, with construction stress-strain curve in the temperature range 20÷1000°C. There were determined transverse fracture stress σ_b , yield stress σ_s , (fig.3), strain to fracture δ , Young modulus E, parameters of strain hardening and ductile-brittle transition temperature.

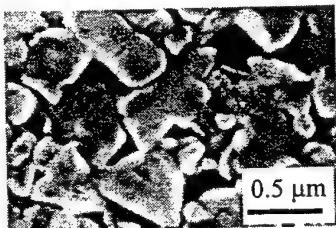


Fig. 1. The structure of alloy
WC-6%Co by mas.,
 $d = 0.33 \mu\text{m}$.

The highest values of hardness 22 GPa were achieved for alloys with submicron grain size, at that dishardening takes place more slowly than for alloys with the micron grain size.

Brittle-ductile transition takes place in WC-Co alloys at heating. The ductile-brittle transition temperature is equal to 200-300°C for micron grades and 300-700°C for submicron grades. But essential plasticity (strain to fracture δ) is observed at temperatures higher than 600-700°C for all alloys. It was shown that very high strain hardening is typical for WC-Co alloys. For alloys with submicron grain size the grain boundary sliding (which is typical for superplasticity) is the leading deformation mechanism at 1000°C.



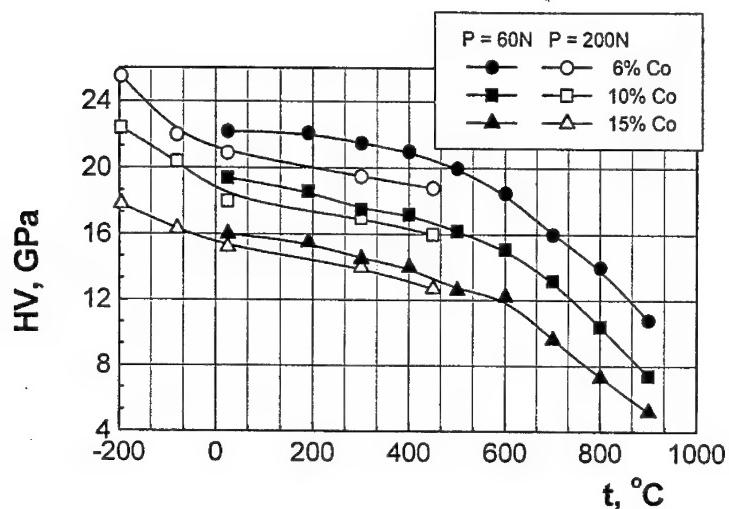


Fig.2. Temperature dependence of hardness for alloys with grain size $d = 0.5 \mu\text{m}$.

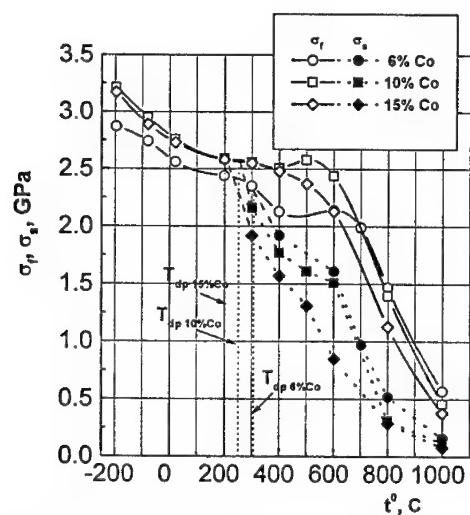


Fig.3. The temperature dependence of the yield stress σ_y and transverse fracture stress σ_f for alloys with grain size $d = 1.17 \mu\text{m}$.
The ductile-brittle transition temperature T_{dp} is shown too.



**EFFECT OF THE NANOCRYSTALLINE POWDER
 ZrO_2 -3 mol. % Y_2O_3 PRODUCTION CONDITIONS ON
 FRACTURE TOUGHNESS OF MATERIAL**

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The materials based on ZrO_2 have a wide functional application. Indentation is used for investigation of fracture toughness of these materials.

The aim of research is to investigate the behaviour of K_{Ic} of materials depend on the variation of the preparation conditions of starting nanocrystalline powder ZrO_2 -3 mol. % Y_2O_3 .

Methods of investigation are X-ray phase analysis, petrography, BET.

Starting nanocrystalline powders ZrO_2 – 3 mol.% Y_2O_3 were produced with high-temperature hydrolysis [1]. Various methods of thermal treatment were used under obtaining of three batches of powders (Z1 - Z3): Z1 – the powder suspensions have sprayed in a hot zone ($t = 200^{\circ}C$) with ultrasound; Z2 –suspensions have annealed under temperatures up to $600^{\circ}C$ without preliminary repeated decantation in a water; Z3 - suspensions have carefully washed from an ion of Cl^- with repeated decantation in a water.

The primary particles size of obtained batches of the Z1- Z3 powders was varied from 5 up to 10 nm, specific surface area- from 20 up to $96\text{ m}^2/\text{g.}$. Powders were agglomerated. The powder Z1 includes the spheres of size from 10 up to 200 μm . The powder Z2 contains the agglomerates to the shape of close to spherical of size up to 5 – 10 μm . The powder Z3 includes the spherical "soft" agglomerates of size up to 10 μm . Zirconia of tetragonal modification (T- ZrO_2) was the phase composition of all three powders.

The tablets from these powders have formed with cold uni-axial pressing and have sintered in vacuum under the temperature of $1450^{\circ}C$ (2 h). The sintering density of the tablets has achieved 98 % from theoretical. The microstructure and phase composition of the samples are various . The phase composition of Z1 and Z3 was T- ZrO_2 and traces of zirconia of cubic modification (F- ZrO_2), Z2 - T- ZrO_2 and F- ZrO_2 (up to 10 %). The microstructure of Z1 does not homogeneous: the T- ZrO_2 single grains (size up to 1 μm) are disseminated in the matrix of T- ZrO_2 (grains of size up to 0,5 μm). The microstructure of Z2 is homogeneous: the F- ZrO_2 grains of extended rounded form (size up to 5 μm) are distributed in the fine-grained T- ZrO_2 matrix (grains of size up to 1-2,5 μm) . Fine-grained microstructure of Z3 is homogeneous with grains of size up to 1 μm .

The critical stress intensity factors (K_{Ic}) of materials are tabulated in the Table.

The critical stress intensity factors of the Z1- Z3 materials are directly connected to their microstructure. The sintering particularities are defined the microstructures of



Table. Critical stress intensity factors (K_{Ic}) of materials.

Materials	P^*, H	H^{**}, MPa	K_{Ic} calculated		
			From [2]	From [3]	From [4]
Z1	60	18811	7,54	7,64	9,88
Z2	100	19813	18,34	14,57	25,36
Z3	100	21000	9,84	8,91	10,29

P^* - indenter loading, H^{**} - microhardness of materials.

the Z1- Z3 materials. Formation of separating connections between the Z1 primary powder particles during their spraying in a hot zone results in the preservation of the granules fragments under brittle fracture during uni-axial pressing. These fragments are become the sources of single larger grains owing to " of zone isolation " under sintering. The absence of repeated decantation under producing of the Z2 powder results in the preservation of impurities of oxides of silicon and aluminium in the powder. The generation of a liquid phase under sintering of the Z2 powder results in growth of larger grains of extended, rounded form located in the fine-grained matrix. Uniform consolidation of "soft" agglomerates under cold uni-axial pressing of the Z3 powder and the absence of braking effect of the chlorine ion under sintering result in the formation of homogeneous fine-grained structure of the material. Enhancement of the microstructural homogeneity (the Z1 and Z3 samples) results in increasing of K_{Ic} in 1,3 times. The formation of the two - scale microstructure (the sample Z3) promotes to increasing of K_{Ic} practically in 2 times. The microstructure variation results in changing of the materials toughening mechanisms. Both transformation toughening and switching of domains are in the Z1 and Z3 materials . Transformation toughening, switching of domains as well as derivation of cracks are in the Z2 materials .

The variation of the preparation conditions of starting nanocrystalline powder $\text{ZrO}_2\text{-3 mol.\%Y}_2\text{O}_3$ give rises to obtaining the different microstructure of materials. Formation of the two - scale structure (larger grains of F- ZrO_2 are located in the T- ZrO_2 fine-grained matrix) causes growth of K_{Ic} of the materials in 2 times.

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STRENGTH STATISTICS AND STRUCTURE OF PSZ

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It seems to be established that strength of brittle solids is determined by size of surface defects and resistance of material to crack propagation [1]. The surface defects can be mechanical injuries or scratches in the case of structureless solids (glasses, single crystals), pores in the case of ceramics, or grain boundaries in the case of solid polycrystalline materials; resistance to crack propagation is characterized by effective surface energy.

It can be derived from above that strength statistics of the materials would be determined by a size statistics of the respective defects. This idea is not original [2], but its experimental checking encounters quite serious technical problems due to necessity in a very large number of specimens for statistically valid estimation of the distribution parameters. In ceramic science where the strength distribution statistics is of most importance due to both large variation in values and low strength values compared with metals, strength distribution uniformity is conventionally characterized by Weibull modulus m .

A method for approximate estimation of the Weibull modulus within a confidence interval using arbitrary number of specimens had been presented in [3]; naturally, the confidence interval converges when increasing the number, in other words, apparent m measurement error diminishes thereat. It had enabled to study dependences of Weibull modulus on various affecting factors [3] and have allowed to verify the interrelation between statistical characteristics of strength and structure of solids in the present work.

A polycrystalline PSZ with composition $ZrO_2 + 3 \text{ mol.\% } Y_2O_3$ was chosen as the object since it was sintered at 2000°C and had therefore zero porosity but quite coarse grains about $30 \mu\text{m}$. Three-point bending specimens about $2 * 3 * 15 \text{ mm}$ in size were cut down from one piece of the material, diamond ground, and annealed at 1400°C . Thermal etching occurred thereat permits easy observation of the grain boundaries even under small magnification. Large grain size have provided the negligible effects of a transformed surface layer which strongly affects the fracture statistics in the fine-grained PSZ overlapping the structure effects [2].

A standard mechanical testing data for 14 specimens were computed using formulae [3]:

$$m = \frac{N \sum_i X_i Y_i - (\sum_i X_i)(\sum_i Y_i)}{N \sum_i X_i^2 - (\sum_i X_i)^2},$$



$$\pm \Delta m = t_\gamma \frac{1}{\sqrt{N(2-N)}} \left\{ \frac{N \sum_i Y_i^2 - (\sum_i Y_i)^2}{N \sum_i X_i^2 - (\sum_i X_i)^2} \right\}^{1/2} \left\{ \frac{\left[N \sum_i X_i Y_i - \sum_i X_i \sum_i Y_i \right]^2}{\left[N \sum_i X_i^2 - (\sum_i X_i)^2 \right] \left[N \sum_i Y_i^2 - (\sum_i Y_i)^2 \right]} - 1 \right\}^{1/2}$$

where

$$X_i = \ln \ell_i, \quad Y_i = \ln[-\ln(1 - \frac{i-0.5}{N})]$$

N is a number of specimens, i is the specimen's current number, σ_i is the specimen's strength, t_γ is a Student's criterion for the confidence probability γ (for technical applications $\gamma = 0.95$). The calculation results are presented in Table 1.

The structural data were obtained from the typical area of etched section about 170 by 170 μm in size contained 50 grains. The largest size of each grain d was recalculated into the parameter $d^{1/2}$ proportional to strength (exact magnitude of a proportional factor plays no role at determination of m), and a 'structural' Weibull modulus was computed using the same formulae. The results are shown in Table 1 as well.

Table 1. Results of statistical calculations

	Mean values	$m \pm \Delta m$	N
Mechanical testing	419 MPa	$4,24 \pm 0,31$	14
Structural data	28 μm	$4,77 \pm 0,36$	50

As can be seen from the table, both Weibull moduli are coincide within their confidence intervals. Two interesting conclusions can be involved from this fact:

a) Both mean values and strength statistics of the brittle polycrystalline materials in the absence of foreign effects are determined by respectively mean values and distribution statistics of grains.

b) Statistical parameters of such materials can be estimated with a sufficient precision from the structure, avoiding labour-consuming and expensive procedures of manufacturing and testing a large batches of specimens.

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NATURE OF THERMOMECHANICAL INSTABILITY OF DEFORMATIONAL TWINS IN DIAMOND AND 3C BN, 3C SiC

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Earlier it was revealed [1] that twinning is main mechanism of plastic deformation in diamond, 3C BN and 3C SiC during thermobaric treatment ($P=8$ GPa, $T>1300$ °C). It occurs in the system of parallel or crossing planes of (111) type. One may consider twins as a packets of stacking faults (10-20 nm). In the common case twins pass through out the crystal.

In the further study it was established that during following deformation of the crystals structure state of twins changes up to the disappearing of twins. For crystals studied it is realized similarly. Distinctions concern initial temperature of reconstruction and destruction of the twins.

It is revealed that reconstruction of twins follows the scheme: violation of coherence in conjugation with matrix → transformation into dislocation volume subgrain boundaries → new structure elements formation (dislocation networks, plane pile-ups, autonomous assemblies, parts of high-angle boundaries). It is interesting that in the same time processes of dislocation ordering do not take place in the matrix.

These features of structure transformations were discussed in view of known ideas about interactions between dislocations and grain boundaries [2]. It was concluded that:

- a) during deformation of crystals with high concentration of defects deformational twins become a sink for dislocations;
- b) rearrangement of dislocations in the twins may be due to occurrence of various reactions and development of diffusion processes in microvolumes;
- c) since dislocation ordering do not takes place in the crystal matrix process of reconstruction of deformational twins may be considered as a sort of dislocation recovery.

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PLASTIC FRAGMENTATION OF 2H BN AND 6H SiC CRYSTALS DURING THERMOBARIC TREATMENT

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Generalization of results of structure transformation study in 2H BN- and 6H SiC-crystals with high concentration of basic stacking faults during P-T-deformation ($P=7,7$ GPa, $T=1200-1600$ °C) is presented. Transmission electron microscopy was used for investigation of thin foils. Crystal sections of (1120) type were studied. Stacking fault lines were considered as marks of basic planes.

On the base of analysis of form changing of plate-like crystals and basic layers in them it was revealed that deformation of crystals with stacking faults realized via transmission of local microvolumes, i.e. by rotary mechanisms. Such deformation determines formation of misorientation boundaries in crystals.

On the view of microstructure features of such deformation one may distinguish 5 individual processes of its development in the crystals:

1. Kinking. It realizes as massive and local via group inclinations of basic layers regarding axes in (0001) plane.
2. Misorientation band formation via local rotation regarding axis [0001] or axes inclined to it.
3. Development of local crystallographic shifts at (1012) and (1012) planes.
4. Occurrence of local crystaloriented rotations regarding of axis [0001].
5. Transmission of microvolumes in direction [0001] or axes inclined to it.

Structure state of boundaries between fragments was identified. On the first stage high-angle boundaries appear at kinking and local rotations only. In other cases there is continuous misorientation. Further evolution of structure state of such boundaries takes place up to the appearance of high-angle misorientation.



NANOCRYSTALS

47-70



DEFORMATION OF $\text{ZrO}_2\text{-Y}_2\text{O}_3$ NANOPOWDERS AT COMPACTION BY HIGH HYDROSTATIC PRESSURE

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The compaction of powders is one of basic operations of ceramics production, which for obtaining of maximum dense ceramics is being often accomplished by the method of hydrostatic pressing. In this case pressure for every individual particle or group of the interacting particles (agglomerates, aggregates) separated by the elastic inpenetrable shell does not hydrostatic (all-around) ones, but only create a load distributed upon shell surface. The work aim is investigation of the deformation processes character which take place at compaction of the tetragonal powders of ZrO_2+3 mol.% Y_2O_3 content at compaction by hydrostatic pressure (HP) in 100-1500 Mpa interval.

The powder main structure element is porous aggregate of 2,2-2 μm consisting of crystalline nanoparticles united between itself by neck and containing about 40% nanosize pores (Fig., a). The powders differed by the particles size was investigated. The investigations were carried out by methods of X-ray structural analysis and electron microscopy. The density was measured by method of hydrostatic weighing

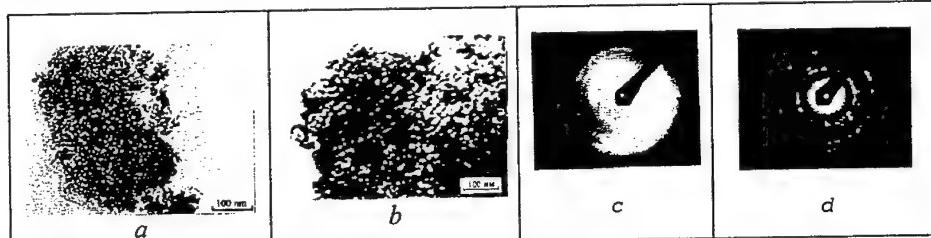


Fig. Structure of ceramics: (a) aggregate, (b) structure of fragment of crushed compact, (c) microdiffraction pattern from fragment, (d) the same after heating.

It was established that at compaction in HP condition at some critical pressure P_{cr} the change of densification mechanism is going on: the aggregates packing is changed by aggregate fragmentation. Value of P_{cr} characterizing the aggregate strength increased with the particle growth that is connected with increase of strength of the interparticles bonds in aggregates. Further pressure increase do not lead to fragments destructurization, as even at the careful crashing of compacts the fragment structure does not change (Fig., b).



At compaction of the tetragonal zirconia powders in HP condition the induced by arised stresses martensitic transformation of tetragonal zirconium dioxide into monoclinic one ($T \rightarrow M$ transformation).

Pressure, P , MPa	Initial Particle size	Density, g/m^3	M-phase quantity	CRS, nm	P_{ck} , MPa
250	10	2,3	24	12,2	410
500		2,66	26	11	
750		2,94	27,5	9,7	
1000		3,23	29	9,4	
250	12	2,37	25	12,5	430
500		2,74	27	11,7	
750		3,04	30,5	10,3	
1000		3,25	33	10,6	
300	17	2,7	26	17,7	500
500		2,97	29	17,1	
750		3,19	32	16,3	
1000		3,38	39	16	

As it seen from table the monoclinic phase quantity increase with the pressure and particle size growth. It was established that $T \rightarrow M$ transformation is connected with the densification mechanisms. As transformed particles are a part of aggregate consisted of rigidly connected particles the volume increase at transformation of tetragonal particles into monoclinic ones ($\Delta V \approx 0,05$) leads to appearance of an elastic stresses. It is reflected by washing away of the diffraction reflexes on microdiffraction patterns (Fig., c). The compact heating, as safe ones so crushed as well, in 600-900°C temperature interval leads to reverse $M \rightarrow T$ transformation and removes intraaggregation stresses. It is marked by disappearance of the microdiffraction reflexes washing out on microdiffraction patterns (Fig., d).

The decrease of coherent scattering region (CRS) at increase of compaction pressure is connecting with $T \rightarrow M$ phase transformation but does no connected with particles destruction as possibility of the destruction of 10-20 nm size particles is being excluded practically at using pressure level at least.

Thus, it is established that deformation of $ZrO_2-Y_2O_3$ powders at compaction by high hydrostatic pressure is going on in account of aggregate fragmentation and martensitic transformation of tetragonal phase into monoclinic one.



EFFECT OF ALLOYING ON METALLOCARBOHEDRENES AND METALL-CARBON NANOCRYSTALLITES STABILITY

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Refractory carbides of transition IV-VIII group *d*-metals (M), which are characterized by extraordinarily wide functional properties, including extreme thermal and strength characteristics, have long been the objects of intense studies.

The amount of known stable chemical compounds in M-C systems (in molecular cluster form) has considerably increased in the last years due to the success in nanoparticles synthesis [1]. It is especially interesting that the content of obtained "molecular" carbides is varied within a wide range, sometimes surpassing the "critical" limit $M/C \geq 1$ known for crystalline carbides MC_x . So, a new class of metal-carbide molecular clusters, the so-called metallocarbohedrenes (metcars), was discovered recently. Ti_8C_{12} ($C/Ti = 1.5$) with a cage-like structure was the first representative of this class. A number of other metal-carbon nanoparticles of different composition and structure have been synthesized later on. The most stable among above mentioned species are the nanocrystallite $M_{13}C_{14}$ and the $M_{13}C_{22}$ cluster, which are decorated with C_2 -dimers. Numerous intermediates of these clusters are known today.

The possibility of synthesis of "high-carbon" transition metal carbides with a unique complex of physicochemical, thermal and strength characteristics from the mentioned above nanoforms is considered. They, undoubtedly, will be determined by cohesion properties of initial metal-carbon nanoparticles. "Alloying" of metcars and nanocrystallites with metal atoms of a different kind may become one of the most efficient methods of predictable characteristics formation.

The subjects of this work are "substituted" nanoclusters based on some titanium-carbon nanoforms: metcar Ti_8C_{12} and nanocrystallite $Ti_{13}C_{14}$. Electron density functional theory was used for theoretical modeling of metal-carbon nanoparticles $Ti_{8-x}M_xC_{12}$ and $Ti_{14-x}M_xC_{13}$ (where M – transition *d*-metals replacing Ti in the molecular carbide frame) and prediction of their electronic, cohesive properties and chemical stability conditions.

We have calculated the electronic properties of a series of 20-atom metcars Ti_7MC_{12} , where all 4*d*-atoms are used as substituting elements. Clusters formed from Ti_8C_{12} (T_h or T_d - symmetry) by substituting M atom for Ti were employed as structural models. The general regularities in changing the electronic spectrum of "mixed" metcars Ti_7MC_{12} have been obtained. They are determined by: 1) a



growth of occupation of states due to an increase in the number of valence electrons; 2) the appearance of new (as compared to the spectrum of the original Ti_8C_{12}) "doping" states. The genesis of the latter ones is associated with: 1) systematical shifting of M-atom states downwards the energy scale; 2) systematical decrease of Md -states splitting into bonding and antibonding states. It is found that atomic interactions in Ti_7MC_{12} depend on the kind and coordination of M-atom. The following changes take place in the $Ti_7YC_{12} \rightarrow Ti_7AgC_{12}$ series: 1) nonmonotonous variation in overlap populations of M-C covalent states with a maximum for Zr, Nb, Mo and a subsequent decrease for the end of the series; 2) gradual decrease of the effective charge on M-atoms; 3) general destabilization of the system related with successive occupation of nonbonding (or antibonding) MO. The calculations performed show that introduction of Y, Zr, Nb into titanium carbohedrene frame can lead to a greater stabilization of the molecular cage. $Ti_6Y_2C_{12}$ clusters may become "superstable" metcars with extreme cohesion characteristics.

Similar studies have been carried out for "alloyed" cubic nanocrystallites. Preliminary analysis allows us to affirm that some "mixed" forms of the nanocrystallite can be "superstable", for example, the neutral cluster $Ti_{10}Sc_4C_{13}$ obtained as a result of replacement of four corner titanium atoms by Sc atoms, or its isoelectronic cations $Ti_{11}Sc_3C_{13}^{1+}$, $Ti_{12}Sc_2C_{13}^{2+}$, etc.

Possible ways of "mixed" metcars and nanocrystallites destruction and the mechanisms of their photofragmentation are discussed on the basis of the obtained data. Calculated electronic energies and strength characteristics are compared with experimental and theoretical data for ternary crystalline carbides – solid solutions $Ti_{1-x}M_xC_y$.

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SHEAR FAILURE MECHANISM UNDER UNIAXIAL COMPRESSION OF NANOSTRUCTURED Ti AT 300 – 4.2 K

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The ductile shear failure phenomenon (by catastrophic sliding-off of one part of the specimen relative to another) has been observed under uniaxial low temperature (at 300-4.2 K) quasi-static compression (with the 0.0004 s⁻¹ strain-rate) of prismatic (2x2x7mm) samples of nanostructured Ti (100 -300 nm average grain size). It was processed by severe plastic deformation through the equal channel angular pressing (ECAP) of commercially pure Ti rods (CP Ti) with the average grain size of 15 μ m.

Previously such phenomenon had been observed only under low temperature uniaxial extension or compression of metallic glass ribbons or rods. It was caused there by local adiabatic heating-up to pre-melting (vitrification) temperatures in the catastrophic shear band preceding the failure.

Investigated nanostructured Ti has yield stresses two times larger than initial CP-Ti. Ultimate plastic deformation was from 34% to 4% (in dependence on the temperature and orientation of samples relative to the ECAP axis). The shear failure phenomenon has been observed at all studied temperatures at the compressive failure stresses ranging from 1.4 to 2.4 GPa.

Shear failure surfaces were oriented nearly at 45 degree to the compression axis. SEM fractography revealed “vein” patterns at all observed shear failure surfaces of nanostructured Ti.

In metallic glasses such a pattern is a consequence of the meniscus instability under rupture of the locally heated (up to the superplastic state) bulk of the band of catastrophic shear, owing to its near sound velocity, small thermal conductivity of metallic glasses, and high value of the failure stress. Similar physical conditions of the shear are realized in our case of the nanostructured Ti.

Therefore a “vein” pattern is considered as the indirect evidence of extreme local adiabatic heating during the shear failure of the nanostructured Ti.

Observed non-uniform space distribution of “veins” and its dependence on the structural state of samples are considered as consequences of a non-uniform internal stress distribution and non-uniform adiabatic heating along the surface of shear and failure.



It is supposed that the micro mechanism of the shear failure of nanostructured Ti is similar to that in metallic glasses and consists in extending of the catastrophic shear by grain boundary dislocations.

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PECULIARITIES OF HARDENING THE CU-N BTI NANOSTRUCTURAL COMPOSITE

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Site effects influence many mechanical and physical properties of nanocrystalline materials (NM), in particular, the character of deformation hardening with the nanostructural state attained by both the methods of severe plastic deformation and the accumulating of plastic deformation due to hydropressing or drawing [1]. Depending on the type of structural components having the nanometric size, one can speak, according to definitions of [2], about the nanostructural, nanophase, and nanocomposite materials. Most of the published articles deal with the first two types of NM which are natural substances. The nanocomposite materials, however, e.g. artificially produced composites if fibrous structure are of special interest since structure components can be designed, controlled and checked. But they have not been enough studied because of difficulties in their production. At the same time such materials are of high practical importance. They can be used in windings of strong pulsed magnets [3] since they may combine a high mechanical strength and high temperature strength with high conductance.

Method of packet hydrostatic extrusion and drawing [1] under a limited diffusive interaction of the components has been used to produce a NM. This is the Cu-NbTi- nanocomposite with structure components of 60...10 nm in size, for which the influence of size factor on the character of its deformation hardening has been investigated.

Criteria have been worked out to estimate the strength of reinforcing fibres and copper matrix of the nanocomposite by differentiating the degree of deformation and hardening of matrix interlayers proceeding on the assumption of additivity of their contribution to composite strength. Those criteria have been used in constructing the dependence of calculated strength of the matrix and fibres on size factor.

High strength of the Cu-NbTi nanocomposite at high degrees of deformation ($\ln R=27...32$) and the fibre size of the order of 10nm is conditioned, mainly, by the influence of structure and size of reinforcing fibres and matrix layers with their small- block structure [4]. Strength characteristics of the investigated NM are described by the Hall-Petch relation, if the fibre size d_f is taken for the characteristics one. On the $\sigma_b(d_f)$ curve there are several portions corresponding to different degrees in interaction between elements and defects of nanocomposite fine structure.



The critical value of the fibre size ($d_f \sim 11.5$ nm) is determined starting from which there is a large deviation from general Hall-Petch relations in the investigated nanocomposite. This violation is explained by the competing deformation hardening associated with the processes of breaking, migration of boundaries of structure components differing from macro- and microdimensional states.

A significant increase in the strength of nanocomposite has been found in the range of its structure-element size of 30...11, nm. The strength of experimental specimens ($N_f=211^3$, $V_f=0.1$, $d_f=11.5$ nm) approximately makes 1200 MPa with a satisfactory plasticity ($\delta=2.0...2.5\%$). Strength of reinforcing niobium-titanium fibres estimated by the rule of mixing has ultimately high value ~ 5200 MPa with the strength of nanocrystalline matrix copper ~ 870 MPa. This is 3 to 9 times as much as the ultimate strength of micro- and macro-dimensional fibres.

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INFLUENCE OF HIGH PRESSURE ON STRUCTURAL MODIFICATIONS IN AUSTENITE OF Fe-Cr-Mn-Ni ALLOY

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The initiation of phase transformation in metastable alloys at a strain under high-pressure promotes shaping of materials with nanostructures.

The metastable state formed by a strain between diamond pyramids with superposition of a shift component. The chamber (Bridgeman's method), was utilized: a) to distort a material immediately in clusters of the testing machine (fig1); b) to conduct structural - kinetic researches immediately in the chamber of a microscope and on a goniometer of a diffractometer.

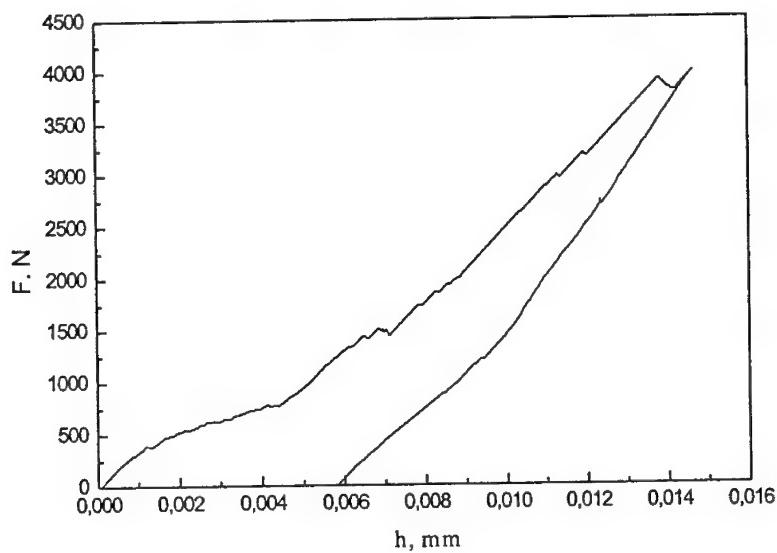


Fig.1 Dependence $F(N)$ on $h(\text{mm})$.

The pressure was evaluated on gain on pyramids and was calibrated on a displacement of R_1 and R_2 lines of a luminescence of a ruby.

The outcomes of experiment have shown, that:



1. The curve of association of force F from depth h of an intrusion indenter had stage character: a) $P \leq 1\text{GPa}$ -stage of elastic behavior; b) $P \leq 10\text{GPa}$ -stage of a plastic strain with fragmentation of elements of structure; c) $P > 10\text{GPa}$ -stage of structural - phase strengthening and secondary elastic behavior.
2. After a strain up to $P \leq 10\text{GPa}$ the alloy remained single-phase (γ -phase) in all range of pressure variation and density of nitrogen (0.06÷0.57%N). By means of roentgen the α -phase did not register, however detected up to 10% of a magnetic phase (in count on α -Fe).
3. The lattice constant monotonically diminished with pressure for all range of density of nitrogen. The widening and intensity of lines (111), (220), (200) is no monotone varied, that correlated with beginning of fragmentation of elements of structure.
4. At pressure $P \leq 20\text{GPa}$ diffuse scattering considerably has increased. Near to a line (111) of the γ -phase detected new, which are identified by a pair of $(110)_\alpha$ and $(110)_\beta$ lines of α -phase.
5. The quantitative handling of diffractogramms and electronogramms has shown presence of elements and complexes with nanocrystall structure.

The formation of α -phase and elements with nanostructure was facilitated at padding superposition of a shift component

It is shown, that the course of a plastic strain at stage of fragmentation and phase ($\gamma \leftrightarrow \alpha$) transformation under high pressure is accompanied by structural - kinetic modifications. It promotes shaping of elements and complexes with nanocrystall structure.

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GRAIN BOUNDARY MECHANISMS OF PLASTIC DEFORMATION OF NANOCRYSTALLINE MATERIALS

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Mechanical properties of nanocrystalline materials (NM) are found to be significantly different from those of the conventional coarse-grained polycrystals. Dislocations are seldom observed in grains of NM and those that are seen have sessile (immobile) configurations [1]. The main structural elements of NM are grain and phase boundaries.

The limited levels of ductility and enhanced strength exhibited by nanocrystalline materials arise of difficulties in creating, multiplying and moving dislocations in nanograins. It has long been observed experimentally in conventional polycrystalline materials that the yield strength varies with grain size through the empirical Hall-Petch relation. Several studies have shown that when NM samples are tested, initial hardening with decreasing grain size is also observed, but that at further reduced grain sizes either hardening at a reduced slope occur or softening is often seen.

A number of models have been proposed to explain this anomalies of grain size dependent strength or hardness variation in NM in terms of dislocations, disclinations and other microstructural features [1].

A model is presented for the initial stage of plastic deformation of NM. It is assumed that low temperature grain boundary sliding may occur on separate parts (facets) by the reorganization of the structural units – grain boundary microsliding (GBMS)[2]. Theoretical study of thermally activated shear transformations in the random grain boundary is carried out. The transformation is in narrow disk shaped region and resemble closely the shear transformations at ingomogeneous plastic flow in amorphous alloys [3]. The parameter τ_s^i must be determined the stress of internal resistance to free plain grain boundary sliding. When an applied shear stress resolved in the plain of grain boundary τ_a exceeded τ_s^i initiated GBMS. τ_s^i -value determined by the grain boundary state and structure and by the temperature.

As sliding proceeds, elastic stresses build up at places where the boundary has curvature, steps or meets other boundaries at triple lines, and the incompatibility at the boundary should be accommodated. The possible accommodated processes are the generation of dislocation or/and the plastic grain rotation. On the basis of a GBMS with accommodation processes model the Hall-Petch dependence anomalies for nanocrystalline materials have been explained and the conditions of its establishment have been defined. The dependence of the yield strength on average grain size breaks up into two regions: the first has an upper bound D^* and is characterized by linear



increasing yield strength with D, while in the second obeys the Hall-Petch relation. Dependence of NM yield strength on the temperature has been calculated. Obtained results are corresponds to some experimental dates.

The volume fraction of the GBMS regions increased with increasing applied stress. The macroscopic plastic flow of NM can be considered as the formation of a percolation cluster of GBMS regions.

Theoretical study of plastic flow in NM with amorphous grain boundaries is carried out. The stress of internal resistance to GBMS related to the strength of an amorphous metallic alloys. The plastic flow kinetics of these NM has been calculated.

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STRUCTURE, STABILITY AND DEFORMATION MECHANISMS OF MATERIALS WITH ULTRA-FINE GRAIN SIZE

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Polycrystalline materials with average grain size in the range of few hundred nanometers are being nowadays successfully produced in a bulk form, which allows investigations of their structure and mechanical properties. Such studies are being intensively carried out in different countries. One of the leading schools in this field is undoubtedly the Russian school. Scientific groups of Moscow, Ufa, Ekanerburg, St'petersburg and Tomsk are successfully working in this area. First review papers on the problems of structure and properties of nanocrystalline materials have been published recently [1-4]. Science is moving very fast in this area. New problems appear and old, unsolved, debatable problems still remain. In this paper, the authors present their opinion of the problems such as nanocrystalline structure, its stability, internal stresses, mechanisms of deformation and deformation resistance.

In the structure of crystals, obtained by severe plastic deformation (SPD), there is an acute problem of the size of real grains, fully surrounded by large-angle boundaries. The sizes indicated in many papers are often underestimated. This question is closely related to the anisotropy of grains, presence of sub-boundaries and structural texture of materials after SPD. Detailed electron microscopy investigations enable to distinguish between three groups of sub-grains (fragments): 1) sub-grains with chaotic and network dislocation structure; 2) sub-grains with cell substructure; 3) sub-grains, free from dislocations. Complex hierarchical structure is organized in such a way that isotropic cells are grouped within anisotropic sub-grains. However, sub-grains and grains are not always organized in the structure of "Matreshka" doll type. Such scheme is related to the mechanisms of structural evolution during SPD. Undoubtedly, an overall picture of transformations in the defect sub-system can not be achieved with the use of only average values of sizes of grains and sub-grains, without detailed description of their types. Statistical analysis of distribution functions of sizes of grains and sub-grains needs to be performed taking into account the fact that all substructural transformations occur at the "tails" of distribution functions. The sub-grains become smaller by means of division by new developing sub-boundaries, and at the same time they get bigger due to the displacement of these boundaries. The intensity of each of these processes depends on the strain rate, temperature and degree of deformation, constraint of deformation and etc.

Internal stress fields, the sources of which are various boundaries, joints of boundaries and second phase particles, play an important role in these processes. The role of the second phase particles are obviously underestimated in the modern studies of the material structure after SPD. The point is that SPD leads to the



intensive redistribution of different impurities in the material volume, especially interstitials and atoms of elements which are almost insoluble in the lattice of main metal. They get captured by dislocations and swept out primarily on the grain boundaries and to the joints of boundaries, where the second phase particles are formed. The presence of particles is closely related to the problem of stability of the structures with ultra-fine grains. The analysis of equations of Ziner, Gladman and Hessold, which relate the size of the stable sub-grains to the volume fraction and sizes of second phase particles, is given in the present paper.

The structure of nanocrystalline materials determines the mechanisms of plastic deformation. With the grain size of hundreds of nanometers, the main deformation mechanisms are intergranular dislocation glide (and microtwinning) and grain boundary sliding. Their relative contributions depend on the type of material, deformation temperature, strain rate and orientation of extended boundaries with respect to the deformation axis. Under conditions of self-coordination of intergranular glide and grain boundary sliding, dislocations are not retained inside the grains, but get pulled in the boundaries. Under such circumstances, the stress-strain dependence becomes distinctive, with the short stage III of parabolic hardening and the extended stage IV with deformation hardening coefficient being almost zero. Such stress-strain curves were first published in [5,6].

The decrease of the grain size normally increases resistance to deformation according to Hall-Petch law. Some complications with validity of this relationship were noticed in the nanostructural range. The analysis of Hall-Petch relationship will be given in the final part of the presentation.

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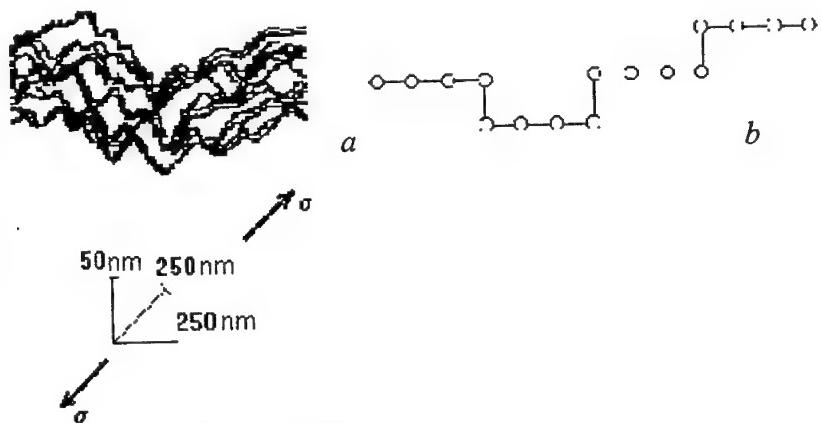
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FRACTAL STRUCTURE OF THE RELIEF OF DEFORMED Ge AND Si SURFACES

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The effect of biaxial lateral stretching on the relief of the Ge(111) surface ($\sigma \sim 0.3$ GPa) has been studied by scanning tunneling microscopy (STM) [1]. It was found that the roughness of the whole scanned surface increased with time under the loading reached some critical level (Fig.1).



*Fig. 1. a) The part of the topogram of loaded surface.
 b) The scheme of atomic terrace.*

We also noticed the self-similar (fractal) character of the relief of the deformed Ge(111) face. That is, with increasing microscope magnification, smaller and smaller relief fluctuations are detected. The scaling of these fluctuations at different levels obey some similarity law. The same effects have been observed on the Si (111) face. However, up to now, our conclusions about the fractal character of the deformed surface were qualitative due to the methodical limitations.

The atomic level resolution for the structure of the relief of the deformed Ge(111) face has been achieved by LEED. We observed anisotropic broadening of reflexes with maximum in azimuth direction $<21>$. This angle broadening has been



interpreted as a manifestation of terrace-like quasiperiodic structure, oriented normal to the azimuth direction <21>.

Geometric parameters of this step-like structure were estimated from the parameters of anisotropic broadening. At the atom-size height of these steps their width was equal to 5 – 7 lattice constants. The length of these terraces was estimated to be about 3 times larger than their width.

Inclination of the surface relief, caused by this step-like structure, is about 10°. This “stepnes” of the surface (the mean slopping being 0.5°) can be realised only via the formation of rough relief.

The structure of this relief on the largest-scale levels, up to the micron level, revealed by STM, is in good agreement with the results obtained by LEED. It should be mentioned that the self-similarity elements in the structures of deformed surfaces were observed on amorph alloys [2] and on polycrystalline metal foils [3].

Acknowledgements.

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EVALUATION OF THE SPECTRAL AND FRACTAL SURFACE STATE OF SOME AMORPHOUS METALS

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Presently the nano- and microscale surface structure of amorphous metals is intensively studied which provides important information for the understanding of a number of mechanical characteristics of these materials [1, 2].

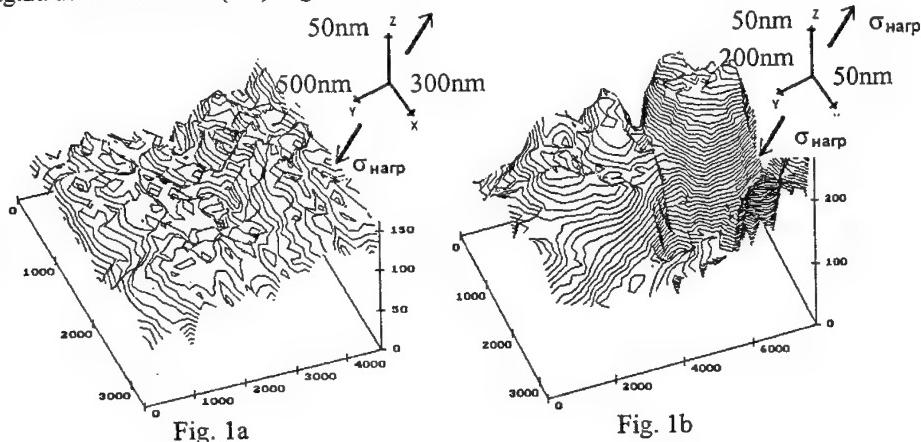
In the present work the foils of Fe and Co-based of melt spun amorphous metals have been investigated.

The aim of our studies was to estimate the spectral and fractal surface state of these metals under tensile stress.

Experimental Images of the surface of the amorphous alloy $\text{Fe}_{77}\text{Ni}_1\text{Si}_9\text{B}_{13}$ subjected to the increasing tensile stress (0 – 3 GPa) at room temperature have been studied by the scanning tunneling microscopy (STM). The data were collected in a digital form. The computer-assisted spectral and fractal analysis of these data was carried out.

Results and discussion Two selected surface images of the same surface area are presented in Fig 1(a, b). In the first case (Fig 1a) the sample was without the stress, in the second case (Fig 1b) the sample was under uniaxial tensile stress of 2Gpa. One can notice the qualitative difference between the images.

Fig.2 shows the dependence of the relative number of defects on their height z Fig.2a and half-width (H_x) Fig.2b. It can be seen that part of the larger defects



consists from a number of smaller defects. It is evident that the presented dependencies strongly differ for "free" and stressed samples.

For the surface images (see Fig. 1) fractal dimensions (d) have been calculated by "Box Accounting method". At $\sigma=0$ $d=2$, while $\sigma=2\text{GPa}$ $d=2.2$.

Previously we observed the formation of fractal surface on the Ge(111) and Si(111) single crystals under tensile stress. The formation of such surfaces took place by developing of steps which resulted in deep and wide valleys and finally, in the destruction of the sample. This phenomenon appears to be general nature.

For a fractal origin the increase of the factuality of the side surface plays the role of a peculiar search for the third dimension (not lateral one). Exactly in this direction the formation of the main crack and the fracture of sample occur.

Conclusions:

Both the spectral and fractal states of the side surface are altered under the action of tensile stress.

The increase of the fractal dimensionality under stress facilitates the origination of the main crack.

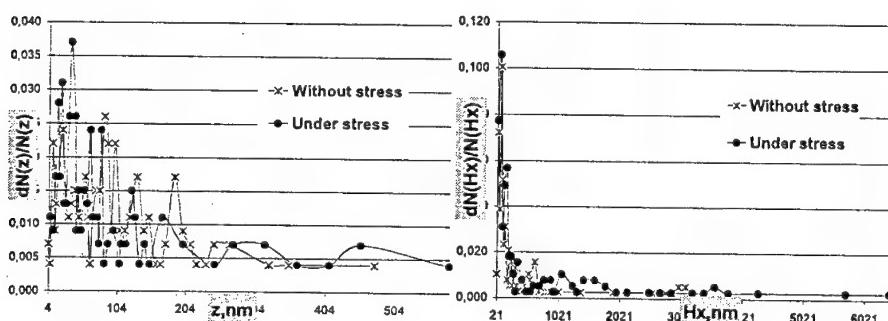


Fig. 2a

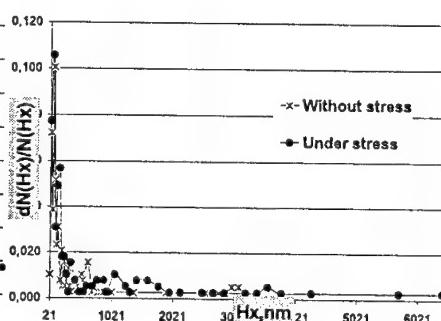


Fig. 2b

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DEFORMATION AND FRACTURE OF BORIDE/NITRIDE NANOCRYSTALLINE MATERIALS

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Unusual properties of nanocrystalline (nanostructured, nanocomposite, nano-size, etc) materials (NMs) which are usually characterized by a grain size in the range of 1-100 nm catalyzed the numerous investigations in this field. However, the nature of the NM deformation and fracture as well as grain boundary structure remains unresolved and needs further considerations. In this connection, fracture surface study by FESEM, HREM investigation and nanoindentation tests seem to be useful and interesting for the better understanding of deformation and fracture nature of NMs.

Nanocrystalline bulks and films based on high-melting point compounds such as borides and nitrides (TiB_2 , TiN , AlN , $Ti(B,N)$, and $(Ti,Al)N$) were prepared by high-pressure sintering and magnetron sputtering and investigated by conventional and high-resolution SEM and TEM as well as by AFM. Strain characteristics have been also studied by nanoindentation tests.

The main obtained results are the following [1 –3]:

- the fracture surfaces are intergranular;
- homogeneous deformation and localized inhomogeneous deformation have been observed in the case of films with clearly-defined columnar structure (TiN , $(Ti,Al)N$, and $Ti(B,N)$) and of those with partly columnar or stonelike structure (TiB_2 and AlN) correspondingly;
- in the case of films with a grain size of 3-15 nm the majority grains are to greatly extent crystalline in nature and have typical clear fringe contrast. Essentially amorphous phases and another ones are absent at the grain boundaries. Only films with very low crystallite size (below 1-2 nm) are difficult to interpret the structure feature but the availability of very fine crystallites (≤ 1 nm) seems to be likely;
- edge dislocations have been observed inside some crystallites. These results favor the estimations [4];
- the of information obtained from nanoindentation tests is not coincide with that from conventional tests.

These results are discussed in detail

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ATOMIC STRUCTURE OF TILT GRAIN BOUNDARIES IN ALLOYS WITH L₁₂ SUPERLATTICE

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Intermetallic compounds and ordered alloys with L₁₂ superlattice have the following important properties: superior strength, oxidation resistance at high temperature, anomalous temperature dependent of yield stress. However employment of intermetallids is limited by their grain boundary (GB) embrittlement. The GB embrittlement depends on alloy, for instance Ni₃Al intergranularly brittle, while Ni₃Fe is relatively ductile. It is necessary to study GB atomic structure for understanding GB embrittlement. A qualitative description of GB atomic structure has been achieved in the framework of the coincidence site lattice (CSL) model. This model is based on the geometric criteria of choice of GB atomic structure. Such approach is not reliable especially for alloys, having several variants of GB structure. Adequate results may be received by parallel using of geometric model and computer simulation taking into account interatomic interaction in the alloy.

In this work we study atomic structure and energy of tilt GBs [100] (0kl) in ordered alloys Ni₃Al and Ni₃Fe. The interatomic interaction was approximated by the semi-empirical Morse pair potentials. The accounts of atomic structure were carried out using of a full atomic relaxation by a method of molecular statics. Energies and atomic structure of steady states of GBs were calculated. Comparison of received results with CSL model was carried out. GBs have several steady states. Atomic structure of these states is not coinciding with model CSL. Possible variants of grain boundary dislocations and their influence on the GB embrittlement are discussed.

Calculation of atomic structure has shown that static atomic displacements along three directions exist in the GB core. The simulations have shown the specific oscillatory of atomic displacements. It is results to GB having a layer structure with alternating zones of compression and expansion. The maximum values of displacements are observed in nearest planes from GB plane. The atoms in GB plane do not shift perpendicular GB. In the neighbouring planes atoms move normal to GB plane in direction from the GB. It led to increasing interplanar spacing and formation excess volume at the GB. The values of excess volume have been defined. Atomic displacement in two other directions (parallel GB plane) have shift of one grain relatively another. Simulations shown that displacement of atoms of different kinds on the same plane have different values, which leads to splitting of biatomic planes into two monoatomic subplanes. The splitting of atomic planes occur to distance $\pm 1.5a$ (a – lattices constant) from plane of GB. The comparison of calculated regularities of GB atomic structure with experimental researches by high-resolution electron microscopy of other authors is carried out.



HIGH STATIC PRESSURE COMPACTION OF AN ULTRAFINE DIAMOND

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During last years a nanostructured material science, which investigates objects of 0,01 – 0,1 micron in size and materials based on such particles has been developed very intensively. This interest is connected with the possibility to realize high physical chemical and mechanical properties which are unattainable when using traditional raw materials.

In connection with these circumstances it's very perspective to use fine diamond powders of 4-10 nm in size by producing new polycrystalline superhard materials (PSHM) with a nanocrystalline structure. PSHM production based on fine diamond enables to make abrasive and cutting tool of high productivity combining high precision and rate of the treatment of ceramics, alloys, plastics and improve working characteristics of the tool itself.

In this work technique connected with the problem of producing nanocrystalline PSHM by compaction of fine diamond under high pressure has been observed. There has been shown that as the result of the compaction there is possibility to get PSHM with dispersed structure and high physical & mechanical properties. Changing the chemical state of the surface and a phase content of compacted particles make it possible for fine diamond particles to be compacted within the wide range of pressures and temperatures. Using this technique we have obtained the materials with the needed working properties.

The analysis of results of PSM synthesis process computer modeling has shown the presence of a considerable distortion in homogeneity of stresses & strains distribution and a presence of a critical zone near which the values of strains and stresses reach the maximum. The "container - compact" system geometry effects on the observed dependence of stresses & strains distribution and PSHM properties that stipulates its optimization. This enables to produce high density-polycrystals under lower pressure.



GRAIN REFINEMENT AND ENHANCEMENT OF PROPERTIES IN TUNGSTEN USING SEVERE PLASTIC DEFORMATION

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The results of the first investigations relating to development and application of severe plastic deformation (SPD) method for refinement of microstructure in hard-to-deform commercial purity tungsten are given in the present paper. Specific features of the SPD process by equal channel angular (ECA) pressing of bulk billets are considered. The microstructure of the billets processed was studied by TEM and X-ray analysis methods. The evolution of microstructure with increasing the number of passes in the process of ECA pressing has been investigated. It has been shown that the given approach enabled to fabricate sufficiently uniform ultrafine-grained structure of tungsten with a mean grain size of less than 1 μm and enhanced properties of the material.



LOCAL ATOMIC STRUCTURE AND SHEAR MODULUS AT GRAIN BOUNDAR IN NANOCRYSTALLINE METALS

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High-resolution transmission electron microscopy (HREM) was used to study the structure of nanophase crystals and in nanocrystalline Cu, Ni, Mo, W.. It was shown that the grain boundar between chemically similar nanograins may have different structures; they may represent crystalline junction (with a transition region of no more than 0.2 nm in width) ; be twin boundaries; or have a more complex structure with dislocations; or even have an quasi-amorphous structure. Interphase interfaces between nanophases with different chemical compositions may represent a transition crystalline layer, or a strongly elastically distorted quasi-amorphous layer of up to 2 nm in thickness or even may have a different chemical composition. When grains of chemically different nanophases are joined, their lattices undergo internal elastic distortions and dislocation dipoles are formed inside nanocrystals.

In situ deformation of nanocrystalline Cu, Ni and nanocrystalline multiphase alloy FeNiNbCuSiB in the column of an electron microscope and activation energy for nanometre-scale grain-boundary sliding and rotation plasticity in Cu, Ni and alloy were studied.

Shear modulus at grain boundar are given for the nanometre-sized polycrystalline Cu.



STAGES AND CHARACTERISTICS OF SUPERPLASTIC DEFORMATION OF A SUBMICROCRYSTAL ALUMINUM - LITHIUM ALLOY

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In the present work the deformation of an aluminum - lithium alloy 1420, having after equichannel angular pressing the size of a grain about $3 \mu\text{m}$, is investigated in the temperature range $T = 320-395^\circ\text{C}$, under tension with constant speed of relative deformation in an interval $10^{-1} - 10^{-3} \text{ s}^{-1}$, and also in a creep regime. It is shown that the axial deformation to the moment of a sample break can exceed 1800 %. For so large values of deformation the treatment of results should be carried out, using true deformations and true stresses. Thus on dependence of true stresses σ_t via true deformations ε_t for a curve of a tension there is a stationary stage, which is described by the formula $d\varepsilon_t/dt \sim \sigma_t^n \exp(-U/kT)$ with constant factors $n \approx 2$ и $U \approx 1 \text{ eV}$, and the deformation rate $d\varepsilon_t/dt$ appears close to the creep rate at comparable stress and deformation. The assumption is stated that the deformation at this stage is described by intergranular sliding, that is typical to superplasticity of materials with fine grains. At the previous stage of hardening $U \approx 1.4 \text{ eV}$, that allows to assume selfdiffusion in volume of grains (dynamic recrystallization at a level of subgrains) responsible for deformation at this stage. The conclusion is made also that in conditions of superplasticity the development of deformation results in formation of equilibrium structure, which remains constant during further deforming and will continue, probably, till specific prebreaking state of a sample comes due to geometrical conditions of sample deforming under tension.

The work is partially supported by Russian Foundation of Fundamental Researches (project № 00-01-00482), the Scientific Council of the International Scientific and Technical Program "Physics of Solid-State Nanostructures (project № 97-3006) and INTAS (project № 1997-1243).



AMORPHOUS METAL ALLOYS

73-88



INVESTIGATION OF STRUCTURAL PARAMETERS OF AMORPHOUS ALLOYS IN CRYSTALLIZATION PROCESS

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The process of crystallization of the amorphous alloys is very complicated. Crystallization may take place in various annealing ranges of temperatures, depending on their chemical contents. The process is strictly accompanied by fragility phenomenon. However, further investigations of the phenomenon require continuous studies.

The main goal of the presented research was initial investigations of the Fe-based amorphous alloy, selected during the process of crystallization. Examined matter was annealed in range of temperatures from 473 to 703 K and examined by X-ray diffraction methods. In this process, the degree of alloy crystallization has been performed.

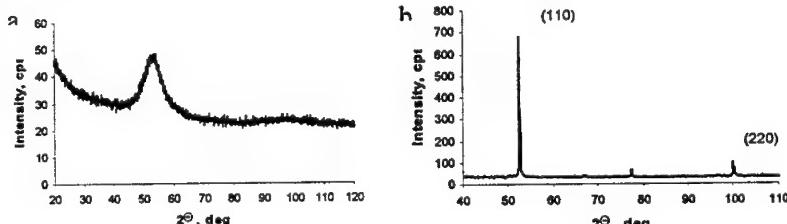


Fig.1. Changes in XRD patterns of amorphous alloy in the process of crystallization by annealing at temperature of: a) 473 K; b) 703 K

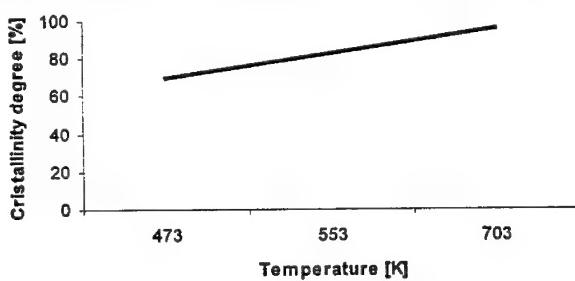


Fig.2. Crystallization process of amorphous alloy at the annealing



On the basis of harmonic analysis methods and changes in shapes of lines 110, have been determined the root-mean-square micro strains ($\langle \varepsilon \rangle$) and coherent block sizes D.

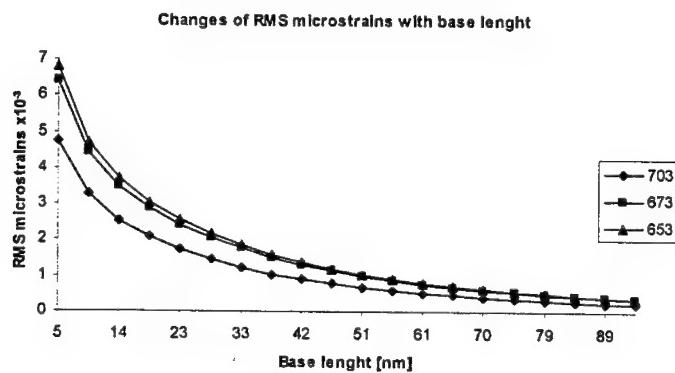


Fig. 3. Changes in RMS micro strains with base length after heat treatment (annealing) in temperature 653 K, 673 K and 703 K

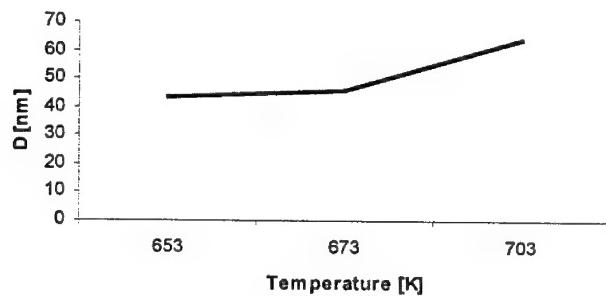


Fig. 4. Changes of size coherent block D after annealing in different temperatures.

Obtained experimental data belong the results of the initial investigations, whose purpose were a precise understanding of the mechanisms and phenomena occurring in the process of amorphous alloys crystallization.

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SUBMICROPOROSITY AND CRYSTALLIZATION OF AMORPHOUS METALLIC ALLOYS

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In spite of the fact that general laws of crystallization of amorphous alloys are well established at present [1], a number of problems and, in particular, specific features of the influence of excess free volume on crystallization still remain unclear.

It was shown earlier that, in addition to the structurally defined free volume which is an intrinsic feature of the amorphous state, amorphous alloys prepared by ultrafast quenching contain elements of excess free volume (EFV) [2,3]. Efv is regarded as structural defects (submicropores with sizes of about 10-100 nm) whose elimination does not lead to changes in the characteristics of the amorphous state. At the same time, evolution of just this mobile component of free volume exerts a considerable influence on physical and mechanical properties of amorphous alloys [3,4].

In this work, the effect of excess free volume on the early stages of bulk and surface crystallization of amorphous alloys was studied by small-angle X-ray scattering (SAXS), electron microscopy, X-ray diffraction, and Auger spectroscopy were used.

Bulk crystallization was studied for the Fe₈₅B₁₅ alloy. The melt was heated before quenching to 1250°C (alloy 1) or 1400°C (alloy 2). Alloy 2 was then cooled to 1250°C. Both alloys were quenched from 1250°C. According to the SAXS data, the average sizes of micropores in both alloys were ≈15 nm, however their concentration in alloy 2 was higher by a factor of ≈1.6 than in alloy 1. In alloy 1, "nonpore" inhomogeneities which can be regarded as quenching nuclei were also observed (in alloy 2 they were nearly absent). It is likely that this fact is responsible for different viscosities of the alloys measured at 588 K: (1.4±0.4)·10¹³ Pa·s for alloy 2 and (8.9±0.6)·10¹³ Pa·s for alloy 1.

According to the electron microscopic data, the average sizes of crystals after the first crystallization stage were nearly equal in both alloys. The sizes ranged from several nanometers to ≈200 nm; however, the surface area of crystals in alloy 2 was larger by a factor of ≈1.25 than in alloy 1. The differing surface areas explain well the experimentally observed difference in thermal effects of crystallization: heat is spent on formation of interfaces, which leads to a lower enthalpy of the first crystallization stage of alloy 2, which was (56±4) J/g, compared with that of alloy 1 equal to (65±3) J/g. At last, the X-ray diffraction data showed that formation of crystalline phases begins in alloy 2 earlier (easier).



Thus it can be concluded that alloy 2 inherited to a certain degree the defect structure of a higher-temperature melt, and the increased porosity of bulk layers of amorphous ribbons facilitated their crystallization.

Surface crystallization was studied for the Fe₅₈Ni₂₀Si₉B₁₃ alloy by SAXS, X-ray diffraction, and Auger spectroscopy. It was found that the onset of crystallization of both the outside and contact sides of the amorphous ribbon occurs earlier than in the bulk. One of the major factors accelerating crystallization at the outside surface was a high concentration of submicropores with sizes \approx 100 nm in the \approx 1-3 μm -thick surface layers. According to spectroscopic data, concentration of basic elements at the outside surface was close to that in the bulk. No specific features in the pore structure were observed at the contact surface; however, it was found to be poor in boron.

Summarizing, it can be concluded that the increase in the inherent submicroporosity of amorphous alloys is one of the major factor facilitating their crystallization upon heat treatment. It can be supposed that changes in elastic energy caused by a decrease in free volume during crystallization are more intense near free surfaces (including surfaces of submicropores) [5].

The work was supported by Russian Foundation for Basic Research (project №99-02-18287)

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PECULIARITIES OF THE MECHANICAL BEHAVIOUR OF AMORPHOUS METALLIC ALLOYS

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Amorphous metallic alloys (AMA) are known during more than forty years. A great many scientific works were devoted to investigation of the structure, properties and formation conditions for these materials since they discovery. In spite of the abundance of data accumulated during whole past period an interest in amorphous metallic alloys don't decrease up to now, that is conditioned by several factors. The main cause due to which AMA still occupy one of the leading places in modern material science is their unique properties. AMA have an extraordinary high strength and hardness with non-zero plasticity, high corrosion resistance and unique magnetic characteristics. Investigation of AMA is also very important for the fundamental science because materials without a long-range order (including amorphous alloys) show in some cases significantly other behaviour than materials with a crystalline structure. It is also known that AMA may be transformed into nanocrystalline or quasicrystalline state through controlled crystallization and are from this point of view the precursors for obtaining the new materials with not less attractive properties.

Amorphous state was realized in many systems including simple binary (Co-P, Pd-Si, Fe-B) and complicated multicomponent alloys (Fe-Cu-Nb-Si-B, Co-Fe-Ni-Si-B, Zr-Al-Ni-Cu-Pd etc.). Necessity of high cooling rates $\sim 10^5$ K/s for amorphous structure formation was the main demerit of previously known AMA. This requirement didn't allow (except for some alloys like Pd-Ni-P, Pt-Ni-P) a manufacture of samples with the thickness more than 100 μm , that along with the complexity of amorphous material consolidation without damage of its structure confined application of amorphous alloys within functional materials. But the new alloy systems discovered at the last years allow formation of bulk amorphous samples with the thickness up to 100 mm [1], that gives possibility to use AMA as the constructional material.

Recently developed aluminum-based amorphous alloys also are of considerable interest due to their mechanical characteristics, which greatly exceed these for the best aluminum alloys with the crystalline structure. For instance, Al₈₇Y₁₀Ni₃ alloy has hardness about 3 GPa and fracture stress of 1100 MPa [2]. Properties of partly crystallized aluminum-based alloys seem absolutely unique. Al₈₃Ni₉Ce₂Fe₁ alloy has hardness higher than 4 GPa and fracture stress of 1560 MPa at plasticity of 2 % [2].

All mentioned facts bring up the question of investigation, explanation and prediction of the mechanical properties of amorphous metallic alloys with new strength.



Peculiarities of the mechanical behaviour of AMA that determined on the base of studying of classical metal-metalloid type amorphous alloys as well as new alloys with high glass forming ability are reviewed in this work. Generalized scheme of the temperature dependence of strength characteristics (H_V , σ_s , σ_f , ε_p) of AMA is presented. It is shown that microhardness and flow stress increase sharply at the temperatures below 300 K, and grows of these characteristics can not be explained by Young's modulus grows.

A whole series of experimental data such as the arrangement of etch pits around the microhardness indentation, slip bands retardation at the intersection of slip systems, value of characteristic temperature of deformation T^* , the character of temperature dependence of hardness and flow stress as well as experimental data available from literature lead to the conclusion that a dislocation mechanism can explain the plastic deformation in AMA in the temperature range of inhomogeneous deformation.

The thermal components of the flow stress and hardness of metallic glasses have value and behaviour analogous to that for crystalline bcc metals. Activation energy and activation volume values that calculated for some AMA applying conception of thermoactivated dislocation moving in field of external stresses are also close to these for BCC metals. At the same time athermal component of the hardness and flow stress of AMA is extremely high.

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MICRO MECHANISMS OF PLASTIC DEFORMATION IN AMORPHOUS METALLIC ALLOYS DURING DYNAMIC INDENTATION

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Amorphous metallic alloys consider more and more practical use. At the same time, frequently enough determining factor in practically use of this materials are playing mechanical properties, particularly, hardness and wear. However is due the fact that what properties and deformation mechanisms of amorphous metallic alloys in conditions of pulse local loading, which frequently realized in practice by impact microcontact wear, abrasive wear, polishing, et cetera. The information about spectrum and dynamics of structural defects and real deformation mechanisms, passing in this conditions, very limits.

Because the objective of this work is kinetics characteristics of indenter penetration determination, active forces and contact stresses and also revelation of mass-transform micromechanisms in material under indenter.

Investigations were performed in a specially developed experimental apparatus for dynamic indentation, with high spatial (20 nm) and temporal (50 μ s) resolutions. At the heart of our methods is placed principle, which include "instantaneous" application of a constant test load to the indenter and continuous recording of the kinetics of its penetration into material with adequate spatial and temporal resolutions. As a result, it becomes feasible to analyze continuously *in situ* the deformation velocity as a function of the instantaneous contact stresses, and to distinguish various phases of the process, and to subject them to thermoactivation analysis. The magnitudes of an activation energy and activation volume allow in turn to understand the micromechanisms of mass transfer under the indenter.

The experiments were carried out on amorphous metallic glass $\text{Co}_{15}\text{Fe}_{70}\text{B}_{15}$, $\text{Co}_{17}\text{Fe}_{68}\text{B}_{15}$, $\text{Co}_{19}\text{Fe}_{66}\text{B}_{15}$, $\text{Co}_{25}\text{Fe}_{60}\text{B}_{15}$, $\text{Co}_{30}\text{Fe}_{55}\text{B}_{15}$, $\text{Co}_{40}\text{Fe}_{45}\text{B}_{15}$ and $\text{Co}_{64}\text{Fe}_{21}\text{B}_{15}$. Well-handled experiments are show, that imprint in all of explored alloys formed by 4-10 ms. The knowledge of the real indenter penetration dynamics affords determine the instantaneous value of velocities; the instantaneous resistance; the dynamic microhardness. The plotting of kinetic curves in semi- logarithmic coordinates allowed to reveal up to several differentially resolved stages in the process of the indenter penetration. For all stages, starting from second, activation parameters of penetration process has been defined. In the second stage the activation volume have been close to 10^{-30} m^3 , at the third - up to $(6-9)*10^{-30} \text{ m}^3$ for all investigated samples.

This means transition from monatomic micromechanism of mass transformation



which in the second stage to correlative movement of low-atoms clusters in the third stage.

Getting dependencies of absolute and relative values of activation volumes at the second and the third stages from percentage contents of cobalt, are shown, that in the region of cobalt concentration 17-19% for revealed stages has extremum in properties, which is reflects on deformation mechanisms.

Thus, well-handled analysis of penetration kinetics, after pulse load, allow us to assert, that the process of local plastic deformation has non-automodelity character, but take place by some replace of each other stages.

We suppose next consecution of mass-transform mechanisms during penetration of amorphous metallic alloys: in the first stage – dominating in all of investigated alloys, apparently, elastic deformation, which replaces on the second stage by monatomic and, finally, on the third stage "quasi-craudions" micromechanisms of mass transformation due to movement of less-atomic clusters.

The value of activation volume and especially it's changing in the transition process from second stage to third more sensitive to percentage concentration of cobalt in alloys, than flow stress and static microhardness of material, that is reflects changing in dimensions and morphology of atomic clusters, that typical for given composition.



**ELASTIC AND ANELASTIC PROPERTIES OF THE BULK METAL
GLASS Zr_{41.2}Ti_{13.8}Cu_{12.5}Ni₁₀Be_{22.5} IN THE TEMPERATURE RANGE 2-20 K**

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Creation of the new perspective class of glass-forming systems viz. multi-component bulk metal glasses stable in broad area of temperatures may be regarded as a noticeable technological achievement of the last few years. The new construction materials, developed on their basis, demonstrate a lot of unusual physical properties, in particular, in the range of low and ultraslow temperatures [1]. It is well known, that the low-temperature properties of amorphous solids can be successfully explained in the terms of a phenomenological model of tunnelling two-level systems (TLS) [2]. It is supposed, that atoms (or groups of atoms) can occupy a number of positions in disordered lattice separated by low potential barriers. At low temperatures the atoms can tunnel through barriers separating their possible positions. In the simplest case, the tunnelling between ground states of the two-level potential is considered and it is supposed, that there is a broad distribution of tunnelling energies and relaxation times. Now it is agreed that the resonant and/or relaxation interactions of tunnelling excitations with phonons and conduction electrons of determine properties of metal glasses in the range $T \leq 2$ K. At higher temperatures a discrepancy between experiment and theory [3] was revealed, and the situation seems not so unambiguous.

In the present work results of an experimental study of acoustic properties of amorphous metallic alloy Zr_{41.2}Ti_{13.8}Cu_{12.5}Ni₁₀Be_{22.5} are presented. Samples had the form of parallelepipeds with dimensions 3 x 3 x 26 mm³. The measurements were carried out using the two-component composite vibrator technique [4].

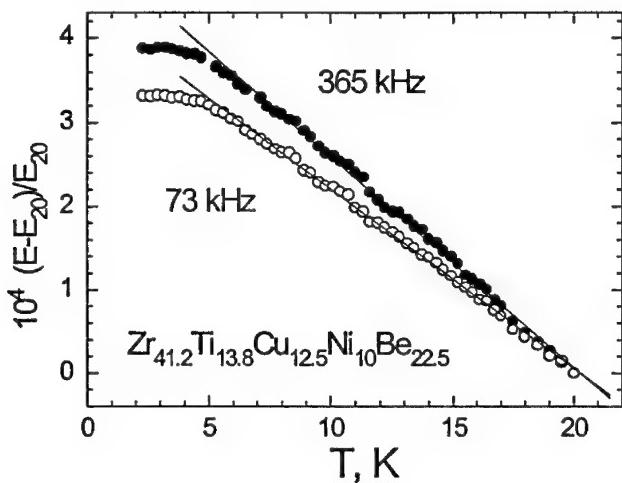
Longitudinal standing waves with frequencies $\frac{\omega}{2\pi} \approx 73$ and 365 kHz were excited in the samples. The measurements were carried out in the temperature interval 2 $\leq T \leq 20$ K.

The temperature dependences of relative change of the Young's modulus measured at two frequencies are presented in the figure. It can be seen that in the interval 5 $\leq T \leq 20$ K the experimental data are described satisfactorily by linear functions $\Delta E/E(T) = A - B T$.

In our experiments $\hbar\omega \ll kT$, and consequently the contribution of resonance mechanisms of TLS interaction with acoustic can be neglected. In the temperature range investigated the relaxation contributions dominate. Ultrasonic phonons disturb equilibrium of the free electron gas and thermal phonons and modulate pa-



rameters of overcoming of potential barriers.



The tunneling model of TLS relaxation predicts the logarithmic dependence $\Delta E/E$ on temperature [5], that is in contradiction with the experiment. The linear run $\Delta E/E (T)$ is rather in agreement with idea about the thermoactivated TLS relaxation [6]. The increase of a slope of the linear part of the temperature dependence of the elastic modulus when increasing oscillation frequency testifies this assumption. According [6] value of a slope B is proportional to $\ln\omega$, that for the frequencies used should cause change in a slope 1.37 times as much. In our case the increase in B was approximately as much as 1.2 times registered. The quantitative discrepancy observed can be explained by a competition of quantum and thermoactivated relaxation mechanisms.

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THE MECHANICAL INSTABILITY OF AMORPHOUS METAL ALLOYS APPEARING UNDER THEIR HYDROGENATION

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The interaction of hydrogen (deuterium) with Fe-, Co- and Ni-based amorphous metal alloys (AMA's) has been investigated. It is shown that the hydrogenation of AMA's in the creep mode under loads constituting 0.1-0.3 of the yield stress activates the deformation of alloys. The deformation stops after the discontinuance of charging with hydrogen (at any stage of creep). Prolonged hydrogenation leads to the destruction of alloys.

It is important to notice that all alloys under investigation remain amorphous after hydrogenation according to their X-ray diffraction patterns. Another feature that can be observed is the decreasing of the shear modulus of the alloys during their hydrogenation. By studying the deformational response of the Finemet-type alloy the new phenomenon unknown before in solid state physics has been discovered. During the charging with hydrogen the alloy loses its carrying power and becomes shapeless with shear modulus value close to zero. This is one of the reasons of mechanical instability of the alloy under applied stress field that is the typical feature of liquids. However it is unusual that the value of its modulus of elasticity doesn't change significantly that is the characteristic of solids. According to these criteria we call the state that appears after hydrogenation of the alloy as quasi-liquid. In 10-15 minutes after charging with hydrogen the alloy reacquires this ability. In 60-80 hours the sample returns to its initial state. Repeating the process of hydrogenation-relaxation several times will cause the sample to destruct. Based on X-ray analysis and X-ray spectroscopy data it was found that the hydrogenation of the alloy changes topological as well as chemical short-range orders.

The appearance of the quasi-liquid state in amorphous alloys under the conditions of structural instability and superequilibrium hydrogen concentration is the process of the liquefaction of AMA that is localized in certain areas of the material. This is one of the reasons of mechanical instability that may be observed during a phase transition occurring in an external stress field.

The hydrogenation of crystalline materials leads to the amorphization of the crystalline matrix but there is no transition to the quasi-liquid state. In case of Fe - based AMA's like Finemet, for which the hydrogen diffusion coefficients are relatively high, the initial absence of short-range order makes the transition to the quasi-liquid state easier.

The authors express gratitude to the Russian Fund of basic Researches for its maintained of works in this direction (grant №99-02-16080).



THE MECHANICAL INSTABILITY OF AMORPHOUS METAL ALLOYS APPEARING UNDER THEIR HYDROGENATION

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**CORRELATION OF THE CRITICAL STRESS FOR SHEAR FAILURE OF
 $\text{Fe}_{85-x}\text{Co}_x\text{B}_{15}$ AMORPHOUS ALLOYS RIBBONS WITH SHEAR BANDS
 CHARACTERISTICS, OBTAINED FROM ACOUSTIC EMISSION
 MEASUREMENTS**

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Plastic deformation under uniaxial tension of amorphous alloys ribbons takes place by nucleation and spreading of localised shear bands, which ensure elementary acts of ribbons macroscopic plastic deformation. A discrete acoustic emission (AE) pulse accompanies appearing of the each band in the amorphous alloy ribbon under strain. Ductile shear failure of amorphous alloys ribbons after plastic deformation realizes by a catastrophic shear at the certain critical stress. This critical stress is one of the macroscopic mechanical characteristics of amorphous metallic alloys.

In this work the complex study is carried out of the critical stress σ_p for a shear failure and of the shear bands characteristics, measured by registration of AE pulses during uniaxial extension of the $\text{Fe}_{85-x}\text{Co}_x\text{B}_{15}$ ($x = 15, 17, 19, 21, 25, 30, 40, 50, 64$) amorphous alloys ribbons. Minima of σ_p had been established in this system at $x = 21$ and 40 at. % Co. These concentrations are stoichiometric in this system.

A complex statistical and spectral analysis of acoustic emission pulses revealed great distinctions of their characteristics in alloys with $x = 21$ and 40 in comparison with alloys of other compositions.

It is established:

- 1) high σ_p and very few acoustic emission pulses prior to failure have alloys with non-stoichiometric compositions;
- 2) critical stresses for appearing of first AE pulses (corresponding to appearing of first shear bands) almost twice smaller for stoichiometric compositions than for others;
- 3) amplitudes of AE pulses as well as their mean full energy are several times higher for stoichiometric compositions than for non-stoichiometric ones. This indicates on more intensive plastic shears in arising shear bands in these cases;
- 4) the full number of AE pulses (correspondingly the number of arising shear bands) is several times more in stoichiometric alloys than in non-stoichiometric ones (latter have only 5-10 shear bands). This indicates that there are more places for shear bands nucleation in alloys of stoichiometric compositions, and these alloys have larger plasticity.



Thus, singularities of mechanical behavior of amorphous alloys of stoichiometric compositions are typical both for macroscopic and for microscopic characteristics of plasticity.

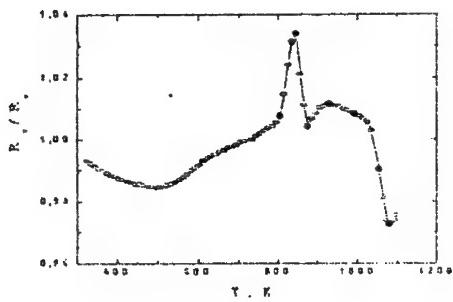
Such a correlation finds its explanation in the framework of the polycluster model of amorphous state. According to this model a nucleation and spreading of local plastic shears – shear bands – realizes along intercluster boundaries. Atomic structure and the shear resistance of these boundaries are different in alloys of the stoichiometric and non-stoichiometric compositions. That is why the correlation is observed between local shear band characteristics (extending to the small distances) and the critical shear stress σ_p for extending catastrophic shear, which spreads through the whole cross-section of ribbons.



TEMPERATURE EVOLUTION THE ELECTRICAL RESISTANCE AND
STRUCTURE DURING THE NANOCRYSTALLIZATION OF THE
AMORPHOUS $\text{Co}_{67}\text{Cr}_7\text{Fe}_4\text{Si}_8\text{B}_{14}$ ALLOY

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Amorphous alloy specimens were prepared by melt-spinning method in Institute of Physics, Sljvac Academy of Science. Variations of the electrical resistance and structure $\text{Co}_{67}\text{Cr}_7\text{Fe}_4\text{Si}_8\text{B}_{14}$ amorphous alloy has been studied during continuous heating at programmed rate of 15 K/min. Electrical resistance R was measured by standard dc four-probe technique. Variation of voltage drop (proportional to ΔR) with temperature was visualized using KSP-4 recorder. X-ray photographs (XSA) were taken on a URS-55 unit in RCU-114 chamber (Co K_{α} -radiation). The temperature dependence of the resistance ratio R_T/R_0 for the studied (R_0 is the resistance at room temperature) is illustrated in figure.



There are four temperature intervals with specific behaviour R_T : 300-773 K, 773-873 K, 873-1000 K and 1000-1123 K. To elucidate the structure variation, the samples were heated at 15 K/min to the present temperature labeled in Fig. (*), water quenched, and characterized by XSA. It was found that the onset temperature $T=843$ K for crystallization coincides with the temperature corresponding to the sharp maximum in the R_T/R_0 curve. The crystallization of the studied alloy proceeds by eutectic mechanism. According to X-ray data, samples heated to above the crystallization temperature, contain a mixture of solid solutions, based on α - $\text{Co}(\text{HCP})$, β - $\text{Co}(\text{FCC})$ and metastable orthorhombic boride Co_3B . In addition, the mixed amorphous-crystalline structure is realized in temperature range 873-923 K in which the dimensions of crystalline phases are 10-20 nm. These dimensions do not alter practically after heating of alloy to 973 K and quantity of amorphous phase is



30 %. A sharp drop R starting at $T=1003$ K are due to the transformation the metastable boride Co_3B into the equilibrium phase Co_2B . It was shown that the X-ray date, the resistance behaviour and the measurements of microhardness H for alloy agree very well in the temperature range 873-1123 K. The maximum $H=10.9$ GPa is observed for the sample after the water quenching at $T=923$ K, minimum $H=7.5$ GPa - after quenching at $T=1123$ K, while for the sample in initial state $H=8.1$ GPa. The temperature $T=497$ K corresponding to local minimum in the R_T/R_0 curve coincides with the Curie temperature [1]. Also, the possible reason of beginning the sharp increase electrical resistance on this curve are discussed.

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MATERIALS WITH HIGH SPECIFIC STRENGTH 90-126



THE STRENGTH AND STRUCTURE OF NANOCRYSTALLINE Ti

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The strength and structure of nanocrystalline Ti (nanograin size $d < 40$ nm) produced by severe plastic deformation were investigated. The data of the electron microscopy research in to the nature of deformation defects are given, and the level of the corresponding strength and plastic characteristics investigated is shown.

Figure 1 show stress-strain diagrams of nanocrystalline Ti with various size of grains tested at 300 K. The yield stress (σ_s) and ultimate tensile strength (σ_B), and relative elongation (δ) presented in Table.

The observation are the results of electron microscopy investigation for the character of defects and natural deformed stripes and the level of the strength leads to the following conclusion: the main places where the deformation of the nanocrystalline material, being investigated in our work, is realized are the grain bounder areas.

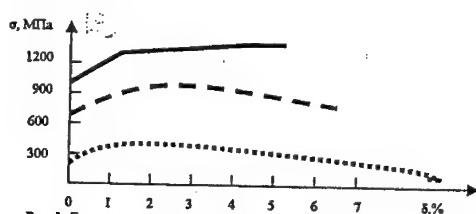


Figure 1. Stress-strain diagrams of the Titanium: ... - $d=100\mu\text{m}$; - - - $d=0.1\mu\text{m}$; — $d=0.04\mu\text{m}$.

Table. The yield stress (σ_s) and ultimate tensile strength (σ_B), and relative elongation (δ), and nanograin size (d).

N	$d, \mu\text{m}$	σ_s, MPa	σ_B, MPa	$\delta, \%$
1	50	275	420	29*
2	0.1	630	645	21*
3	0.15	-	730	18*
4	Деф.50%	510	700	6.5
5	0.04	980	1310	4.5

* - The results of S. Malyshova.



**THE STRUCTURE AND MICRO HARDNESS OF ALUMINIUM
PRODUCED BY RAPID QUENCHING OF A MELT**

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The aluminium rod was produced by rapid quenching of a melt and deformation (70%). In the resulting aluminium had a submicrocrystalline structure with a grain size of 0.5.... 2.0 μm (Fig.1 a,b). The structure aluminium rod with a grain size of 0.5....1.0 μm former upon deformation is characterized by strong distortion of the boundaries of grains.

The microhardness of submicro-grained aluminium is 0.6 GPa. It is the strong microhardness. The submicrocrystalline-copper (a grain size of 1.5 μm) have the microhardness 0.5 GPa.

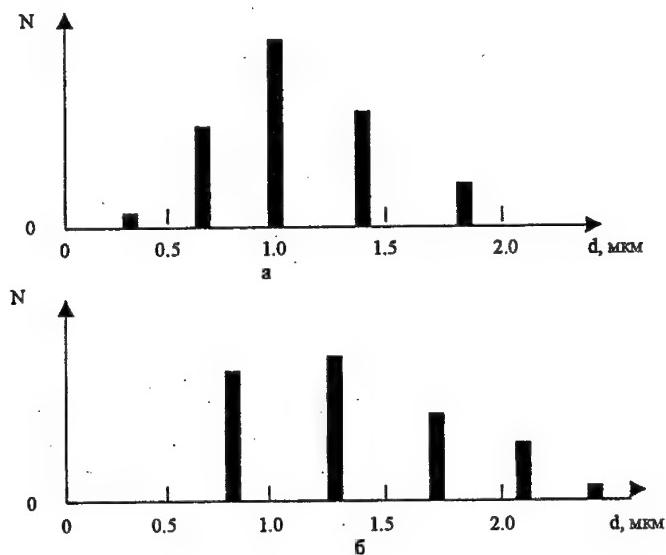


Figure 1. Gistograms of a grain size of the submicro-grained aluminium rod (a diameter -7 mm): a- cross-ways the axle of a rod; b- along the axle of a rod.



RECOVERY OF PHYSICAL PROPERTIES OF 45G17 STEEL BY TREATMENT OF ELECTRIC CURRENT

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The problem of a fatigue is one of the relevant problems presently, as the large number of parts of machines is subjected to a fatigue failure. In time to notify and to stop undesirable destruction, to increase working resource is relevant tasks at today. One of ways of increase of fatigue resource is the effect by electric current impulses on items working in modes of fatigue loads.

In activity of research were conducted with samples from 45G17 steel ($C \leq 0.49\%$, $Mn < 19\%$, $S < 0.3\%$, $Si < 0.1\%$, $Cr < 0.5\%$) with a stress raiser by the way of semicircular cut-out. As a non-destructive technique of accumulation of microdamages the measurement of speed of ultrasound in investigated samples was used. These researches have shown, that after loading the speed of a ultrasound decreases on 0,13 %, and the microhardness thus increases with 2270 MPa in a reset state up to 2430 MPa after loading (load on indentor 100g, time of loading 10s). But already after processing by electric current impulses ($f = 70\text{Hz}$, $t = 20\text{s}$) the speed of a ultrasound increases on 0,03 % as contrasted to by loaded condition, and the microhardness, to the contrary, decreases with 2430 up to 2360MPa after electrostimulating. Though thus the values of both parameters do not reach initial. At processing by electric current impulses is model with макротрещинами, to the contrary, it is possible to watch increase of a microhardness.

In an observed data of speed of a ultrasound during fatigue loading is detected, that the curve of relation of speed of a ultrasound from number of cycles of loading has a kind of a three-stage waning curve. At processing by current impulses at the third stage of the given relation is established, that fatigue resource, which one samples maintained before destruction is increased on the average on 23%. Besides at an electrical stimulation there is an increase of frequency of ultrasound impulses from $36250 \pm 25\text{ Hz}$ after loading up to $36296 \pm 27\text{ Hz}$ after electrostimulation.

Thus, on visible the electrical stimulation promotes the stress relief which has collected during fatigue loads, in a corollary that there is an increase of fatigue resource, speed of a ultrasound and reduction of a microhardness.



MECHANISM OF ELECTROSTIMULATED INCREASE OF STAINLESS STEEL X18H10T RESOURCE IN LOW-CYCLE FATIGUE

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Action by powerful short current impulses with definite parameters on frequency, current density and continuity of impulses in a definite area of curve of low-cycle fatigue S-N allows to increase the products resource from different steels by 15-30% [1].

Nature of reduction effect of fatigue strength is connected mainly with treatment of sub- and microcracks in current action and besides the role of electrostimulation with evolution of defect structure and phase composition is taken into account. In this work such way was used for stainless steel X12H10T. All modes of mechanical loading and electric stimulation were similar to described ones in [1]. Investigation of dislocation substructure showed that in electrostimulated specimen during the process of fatigue tests the structures being different qualitatively on unstimulated specimen are being formed. Cellular substructure is present in small quantities (0,15) and mainly (0,85) the cell-net one. In case of electrostimulation the character of substructure construction is of really low-finished character in comparison with unstimulated specimen.

In finished dislocated structures, as a rule, there are the boundaries of division (in our case they are the boundaries of cells and fragments). The nucleus of microcracks can be formed and spread according to them, especially by deoriented boundaries of fragments. In cell-net substructure there are no such boundaries of division. Therefore, probability of microcracks formation in it and consequently in electrostimulated specimen as a whole is near to zero.

Microdeformation analysis of steel structure revealed as in initial specimen (unstimulated one) the presence of ϵ - martensite in zone of fracture. From analysis of result it follows that the electrostimulation due to the development of second slipping suppresses the process $\gamma \Rightarrow \epsilon$ of martensite transformation. It is necessary to note that martensite transformation occurring in the process of deformation strengthens steel, increasing at the same time its brittleness. Suppression of $\gamma \rightarrow \epsilon$ transformation so plastifies the material. Because in electrostimulated one ϵ -martensite is in small quantities and on the background of cell-net dislocation substructure, it is quite evidently that the stability to nucleation of microcracks must be higher than on qualitative level in analysis of the type of forming substructure.

Quantitative analysis of specimens structure showed that the parameters change characterizing the state of dislocation substructure (scalar density of dislocations



$\langle\rho\rangle$, the curvature – torsion of crystalline lattice χ , the density of curved extinction contours ρ contr. Namely: electrostimulated specimen is broken in higher values of scalar density of dislocations and density of curved extinction contours, the small quantity of microcracks and the small volume of curvature-torsion of crystalline lattice in comparison with initial specimen. So, one can make a conclusion that the electrostimulation plastifies the material in fatigue stress. This effect is in the possibility after electrostimulation to accumulate the large scalar density of dislocations before the microcracks begin to be developed. Picture of internal stresses after electrostimulation shows the large component of curvature-torsion of the crystalline lattice in their low amplitude and, consequently, the less density of microcracks. Carrying out the metallographical and electro-microscopic researches of steel 12X18H10T structure, undergone by low-cycle fatigue test showed that the plastified effect of electrostimulation has a multifactorial character and is first, in occurring the processes of collecting recrystallization, second, in changing the kinetics of selforganization of dislocation substructure, third, in initiating the decay of solid solution with extraction of particles of carbide titanium and, fourth, in suppressing the martensite $\gamma \Rightarrow \epsilon$ deformed transformation, fifth, in developing of defective structure because of deformation transformation and return. The action of electrostimulation on local processes in dislocation substructure, bringing to acceleration or breaking of dislocations gains a special importance. Because of this the electrostimulation breaks the evolution of dislocation substructure and, especially, abruptly decelerates the development of dislocation boundaries of deformation origin. Simultaneously $\gamma \rightarrow \epsilon$ of transformation decelerates. The total of all number of processes are the sharp decrease of density of possible places in nucleation of microcracks in electrostimulated austenite steel and its large plastification in conditions of fatigue loading. All number of these processes trouble the nucleation and development of microcracks moving the fracture to higher number of loading cycles.

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THE INFORMATIONAL MODEL OF THE DEFORMED POLYCHRYSTALIC MATERIAL

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The goal of the study is to reveal interconnections between structural and informational characteristics of the deformed polychrystallic material on the mesolevel. The physics of the plastic deformation processes in mesomechanics has been always proposing the existence of the mechanisms of the primary and accommodation co-action between the rotation and translation modes of deformation.

1. **The model's formalization.** The presentation of model material as an informational system is based on the conceptual idea of the existence of matter and energy as a derivative product emerging in the course of the process of information transmission, accumulation, and processing [1]. At the same time, the deformed material itself pertains to the open, non-linear dynamic, multi-level hierarchical, dissipative self-organized systems. The model is based on the description of the non-linear co-actions of large-scale deformation defects called mesodefектs. The evolution of the mobile mesodefектs leads to the emergence of dissipative structures of the correlation type, as a rule. Functional ordering also furthers the emergence of the geometric structures characteristic of the mesolevel [2].

The deformed polychrystallic material on the mesolevel is represented as the multitude U of its structural levels and the multitude R of connections (relations) between them. Both U and R multitudes called the informational structure SS , are divided on the lowest-level infra-systems, the dynamics of which is defined by the co-actions of mesodefектs in representative scope. At the same time, in every next level, the systems of the previous level are mesodefектs, that is, mesodefектs' level distribution is reflected by the system of enclosed multitudes:

$$U_j \rightarrow U_{j+1} \dots \rightarrow SS \quad (1)$$

2. **Model's description.** We conceptually emphasize concerned and translated information. Concerned or structural information $C(S)$ on the deformed material is defined by the aggregate of statistical data on each structural level's mesodefектs (1). The concerned information is a measure of the structural complexity of material under deformation.

$$C(S) = -(f_1 \lg f_1 + f_2 \lg f_2 + \dots + f_n \lg f_n) = -\sum f_k \lg f_k \quad (2)$$

f_k – the parameters of the frequency distribution of deformation mode for the mesodefектs of the calculated structural level.



Translated information $I(S)$ is represented as reflection of concerned information of the lower level on the structure of the upper hierarchical level, and defined as conditional information.

$$I(S) = -[\sum f_k(E) \lg f_k(E) - \sum f_{k\omega}(E) \lg f_{k\omega}(E) - \sum f_{ky}(E) \lg f_{ky}(E)] \quad (3)$$

$\sum f_k(E) \lg f_k(E)$ - concerned information of the calculated structural level defined by linear plastic deformation.

$\sum f_{k\omega}(E) \lg f_{k\omega}(E), \sum f_{ky}(E) \lg f_{ky}(E)$ - conditional information depended on the realization of deformation's 'displacement – rotation' mechanisms.

3. The experimental basis of the modelling. The geometric image of a mesodefect on the lower structural level is the cell of the divisible grid of 10 mk. We identify three structural levels [2]. The structural levels of larger size are characterized by a range of primary cells. We study divisible grids on the 10 mm length of model material (copper, M0) under static deformation condition and different testing conditions (material's grain size, load speed, sample's working length, rigidity of the experimental machine). The bias field is transformed into distortion field by analytical differentiation $\beta_{ij} = \varepsilon_{ij} + \omega_{ij}$. For the two-dimensional case, Coshi's formulas are employed to calculate all components of β_{ij} tensors of symmetrical and asymmetrical composing parts.

4. Model's investigation. The space-time analysis of the model shows that the change of concerned and conditional informations has the sight of relaxation oscillations. This suggests that the deformed polycrystalline material, as an open system, adapts to the deformation conditions by developing deformation structures, and returns into its stationary condition because of entropy outflow and information inflow on each of the structural level.

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ON THE LOCALIZATION MECHANISM IN Al-Zn-Mg ALLOY

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The literature contains a number of models describing the onset of strain localization but little works which describe experimental investigations of the microstructure transformations which accompany localization. In the present study metallographic and transmission electron methods have been used in order to observe the development of localization in Al-Zn-Mg alloy.

The Al-6%Zn-3%Mg alloy containing the microadditions of the transitional metals is the complex multicomponent system. Specimens were deformed at various strains by compression and tension. Before deformation specimens were quenched from 633 K and then aged (some specimens were only quenched). Such kinds of heat treatment were used to give a supersaturated solution (quenching) and incoherent η phase (aging).

The deformation surface relief was studied by means of optical and transmission electron microscopy (TEM) methods. For the TEM observations two kinds of the shaded replica were prepared for continuous and broken deformations (experiments with repolishing). There was obtained a wide variety of the slip line characteristics for the aged and quenched alloy.

The results obtained for optical investigations of the aged alloy show that the first fine slip lines are observed after compression deformation of 0,02. The quota of the surface covered with such slip lines is small (~0,05) but after compression deformation of 0,015 this quota is already equal to 1. The coarse slip lines formation starts after compression deformation of 0,07. The number of such lines increases rapidly and if the deformation is more than 0,3 the surface of the specimens is completely covered with the coarse lines. The results of TEM observations show that the slip line characteristics are almost independent on strain. Such lines have the average length about 0,3 μm and shear power about 5-7 nm. The dependencies of the characteristics of the coarse slip lines on strain are rather strong and not monotonous. The average length of such lines is 5-8 μm for various deformations, average shear power changes between 60 and 160 nm.

The coarse slip lines were not detected if the second type of the replica was examined (experiments with repolishing). However the new kind of the slip lines was observed by means of these experiments. Such slip lines have the length which is comparable with the length of the coarse slip lines but their shear power is comparable with the power of the fine slip lines. This kind of the slip lines was called the fine long slip lines. The conclusion has been made that if deformation is continuous every "old" slip line observed on the surface of the aged alloy is



agglomerate of the slip lines which lie close to each other. It should be noted that slip line is observed if the shear zone cross the surface of the specimen.

Thus, the results obtained for the aged alloy show that two types of the shear zone called "weak" and "powerful" are formatted in such alloy. The new shear zones as "weak" as "powerful" are formatted close by "old" zones, therefor the majority of the slip lines is the slip bands in fact.

The results obtained for quenched alloy show that if the deformation increases every "old" slip line becomes gradually more coarse. Thus, the main shear localization mechanism in this alloy is activation of the "old" shear zones.

The deformation relief evolution of the aged and quenched alloy subjected to tension tests has the other regularities by comparison with compression tests. The most important distinction: the coarse slip lines are not observed after tension deformation although the aged specimens were broken at strains about 0,05-0,07.

The results of the dislocation substructure (DSS) investigations show that at low strains the net dislocation substructure occupies the basic volume of the age alloy. After compression deformation of 0,15-0,17 the substructure with the multidimensional misorientation is most typical although the volume fraction of the net dislocation substructure is rather great. After compression deformation of 0,02 the block substructure of two types was detected. The first type of the blocks was observed at low strains. The average size of such blocks is equal to 225 nm. The second type of blocks was observed at more high strains ($0,07 < \epsilon < 0,25$). Essential misorientation were detected between such blocks which average size is about $1\mu\text{m}$. The local density of the incoherent precipitates for such regions as measurements showed is smaller than average density.

The regularities of DSS evolution of aged and quenched alloy are essential different. The main distinction is the absence of the block structures in the quenched alloy. It should be mentioned that such structures are not observed after tension deformation too.

The block substructures (fragmentary substructures) are detected in the Al-6%Zn-3%Mg alloy after high-rate shock loading if the deformation localization bands are formatted. It has been found that the block structure formation starts after compression deformation of 0,07-0,09, and the first coarse slip lines make appearance on the surface of the aged alloy after deformation 0,07. Moreover the dependencies of the average characteristics of the coarse slip lines on the strain are similar the dependence of the volume quota of the block structure on the strain. Thus, the results of observation seem to be consisting with such suggestion: the shear zone formation leads to the local substructure transformations. The results of such transformations are the formation of the local softened regions in aged alloy and formation of the local hardened regions in the quenched alloy.



THE DISLOKATION STRUCTURE EVOLUTION OF THE HIGH STRENGTH ALUMINIUM ALLOY BY SHOCK AND STATIC LOADING

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Evolution of the dislocation structure in Al-Zn-Mg age-hardening alloy after high-rate shock loading is studied by transmission electron microscopy on thin foils. Composition of the investigated alloy corresponds to high strength state for the given class of materials.

Striker has rate $V=900\text{m/s}$ in the touch moment with Al-Zn-Mg alloy surface, then it is introduced on some depth and stopped.

The investigated foils were cut out in the zones arranged at different depths along the surface of the channel formed by shock. A-zone is near shock surface on about 1mm depth, B-zone is on the level of the middle part of the channel, C-zone is on the level of the channel bottom. Investigating the dislocation structure in different areas of the dynamically loaded Al-Zn-Mg alloy one can see various amounts of all types of the substructures which are typical to pure metals and monophase alloys with a stable lattice. In the upper (A) and in the middle (B) zones one can see mostly two types of substructures: the net dislocation substructure with smooth misorientation (extinctial contours) and the substructure with multidimention discrete misorientation (orientation chaos). A quota of orientation chaos is smoothly increased with inculcation of the deforming body into the investigated material, reaching its maximum in B-zone and then it is decreased. A quota of the dislocation substructure with smooth misorientation is decreased from A-zone to B-zone and then it is increased. In the low C-zone a quota of contours and chaos turned to be almost equal. Besides, a quota of the net dislocation substructure without misorientation is increased. Fragmentary substructure is another type of the investigated substructures where one can observe any fragments separated from one another by a visible discrete boundary of any orientation both by lowangly boundary and highangly one. A quota of the fragments is smoothly decreased from A-zone to C-zone. The other dislocation substructures occupy insignificant volume.

Moreover the dislocation structure has been investigated at the (1-2mm) distance from the walls of the channel made by the striker. Two types of the dislocation substructure occupy the basic volume along the channel surface of the deformed Al-Zn-Mg alloy: 1) orientation chaos, 2) the net dislocation substructure with smooth misorientation. A quota of the orientation chaos is smoothly increased to the channel bottom (up to 92%), than it is decreased practically up to zero. A quota of the smooth misorientation is decreased with the introduction of the striker, than it is increased sharp occupying the whole volume near the channel bottom.



The fragmentary low disorientated ($1,5-3^0$) substructure quota $P_V=40\%$ has been founded under the striker on the depth 1-2mm.

The Al-Zn-Mg age-hardening alloy dislocation structure after low-rate loading by compression and tension is studded in this work too. Various quantitative characteristics of dislocation substructure were defined by compression tests. In particular volume quota of the various types substructures is defined depending on the degree of deformation (ϵ). It is founded that the net dislocation substructure without misorientation occupies the most part of the volume at the deformation $\epsilon < 0,2$. The such substructure non-uniformity is increased at the deformation increasing, and the volume quota is decreased. Misorientation substructures (with smoothly and discrete misorientation) occupy the most part of the volume of material at more degrees of deformation ($\epsilon \geq 0,2$). Similar orientation chaos substructures are observed at deformation $\epsilon \geq 0,3$.

Block (or fragmentary) substructure which appears in Al-Zn-Mg alloy after low-rate loading by compression is in the appreciable amount too. Two kinds blocks are founded. First kind blocks are observed at deformation interval $0,03 \leq \epsilon < 0,1$. The shape of such blocks is approximately equiaxial. There size are small (average size is 225nm), blocks misorientation is insignificant. Incoherent strengthening particles are on boundaries of blocks. Volume quota of first kind blocks approaches 20% at minimum deformation and it decreases with ϵ increasing. Second kind blocks are founded in deformation interval $0,08 \leq \epsilon < 0,25$. The such substructure maximum volume quota (40%) is observed at $\epsilon = 0,15-0,17$. The misorientation of second kind blocks is more then the first one. The average size is 1 μ m. The block shape is essentially inequiaxial. Incoherent particles are not observed on block boundaries, they are inside blocks. It is necessary to note that such blocks are not observed separately. As a rule they occupy areas with size of the order of several μ m. The blocks (fragmentary) substructure is not observed at the deformation $\epsilon \geq 0,25$.

The deformation relief investigation shows that coarse slip lines appear on the surface of specimens at the same deformation that second kind blocks inside material.

Investigated alloy specimens are destroyed at low degrees of deformation ($\epsilon \leq 0,06$) by the low-rate tension loading. Only net substructure without misorientation with different dislocation density is founded at the appointed interval deformation. Coarse slip lines are not observed at the deformation relief investigation.



USING THE LAWS OF FRACTURE PROCESS UNDER ESTIMATION OF STRENGTH RELIABILITY OF WARES FROM POLYMERIC COMPOSITE MATERIALS

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It is well known, that the wares from polymeric composite materials (PCM) have high weight efficiency that in many respects predetermines their use in flight vehicles. At the same time, sometimes there is a necessity to ensure not only the highest (with allowance for all other rational requirements), but also strictly predetermined strength and load-bearing capability levels.

As a result of researches directed to development of wares strength diagnostics technology, the acoustic emission criteria permitting to determine the actual strength "in-situ" under small test influence were developed. Besides, use of the certain parameters of acoustic emission has allowed developing the approaches permitting to predict the properties of wares from polymeric composite materials already at their designing. These approaches are based on traditional use of four strength reliability models - material, shape, loading and fracture. However essential modifications are introduced into models of material and fracture. For example, the model of a material is usually considered as a model of a continuous medium, which does not correspond to real behavior of any material, especially composite one. And the fracture models are considered a little mechanistically.

Thus, in the offered approach to strength evaluation of elements of constructions, a major moment is the preliminary determination of dynamic change of a state of local elements with taking into account the possible damages.

The methods to determine dynamic change of mechanical behavior of materials should be selected with allowance for what kind parameters being determined and possibilities of determination methods themselves.

So, the determination algorithm for strength of a specimen or ware can look as follows:

- a) determination of dynamic change of necessary parameters of mechanical states;
- b) creation of the model of a material as sets of arrays of structural elements with appropriate parameters of mechanical states;
- c) choice, proceeding from shape and loading models, of the "loading step" magnitude. The "loading step" represents a change of selected state parameters under the change of the loading factor, for example, the magnitude of a load under static loading, time interval under cyclic loading, etc;



d) computer "loading" of the model of a specimen or ware with taking into account the appearance of stress concentration due to arising from fractures of structural elements;

e) determination of strength (load-bearing capability) after reaching by a defining mechanical state of the certain level. Such criterion can be reaching of a limiting state on certain percent of destroyed elements, full fracture of a layer, etc.

The use of new conceptions in models of strength reliability has allowed to approach to designing PCM wares from other points of view. The model of a material is formed within the framework of computer modeling on the basis of tests of separate layers of PCM which are usually tested in two perpendicular directions (for example, in case of an unidirectional composite, the tests are carried out along and across fibers). Then the package of a material "is assembled", the elastic characteristics are determined, and taking into account the shape of an ware and character of its loading, the "loading step" is determined. Each "step" is accompanied by modification of a state of material's model with taking into account fracture model, and the number of "steps" determines ware strength.

The detailed procedure of realization of strength calculation and optimal designing of wares from polymeric composite materials given in the report.



“SCANDIUM EFFECT OF STRENGTH INCREASE
IN WROUGHT WELDABLE ALLOYS BASED ON Al-Mg-Mn SYSTEM

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Among alloys with high specific strength aluminum alloys hold a leading place. If the ratio of the yield stress σ_y (MPa) to the density ρ (g/cm^3) for a constructional steel 40Kh is 130, for high-strength alloys Al-Zn-Mg-Cu it exceeds 200. Al alloys are specially effective under a continuous load, where the specific strength is characterized by the value $\sigma_y^{1/2}/\rho$.

Alloys on the base of the Al-Mg system with Mg content to 6 wt.% (magnaliums) are well wrought and weldable, have high ductility and corrosion resistance, but σ_y/ρ for them even in a work-hardened state does not exceed 160-170. The best way for the additional hardening of magnaliums is alloying by scandium.

A diverse effect of scandium on improving the properties of aluminum alloys is explained by unique properties of the intermetallic Al_3Sc , which has an $\text{L}1_2$ lattice with the parameter exceeding very weakly (by 1.37%) the lattice parameter of Al. Primary Al_3Sc particles, which precipitate while crystallization from the melt at scandium content larger than 0.5 wt %, are crystallization centers and transform ingot structure from dendrite to polyhedral one, which decreases ingot chemical inhomogeneity and sharply improves its workability, and creates a fine fiber structure in wrought semi-finished products as well. Secondary Al_3Sc particles are completely coherent with the matrix, very disperse and very slowly grow with the increase of the temperature and holding time. These particles cause a disperse hardening and a delay of recrystallization beginning. Eutectic character of crystallization and small crystallization interval (~ 5 °C) allow to produce strongly supersaturated solid solutions of Sc in Al by increased cooling rates. By the addition of Sc to the welding wire (to 0.5%) a fine-grained non-dendrite structure of the welded joint is created and the recrystallization in the zone close to the joint that essentially increases the strength of welded joints, especially after annealing.

The Sc effect significantly increases in the presence of small additions (to 0.15%) of Zr, which dissolves in Al_3Sc with slowing the rate of particle growth and decreasing the necessary amount of Sc: thus, in the presence of Zr the ingot structure is modified by alloying with ~0.2% Sc.

Al-Mg alloys are specially favourable for alloying by Sc, because Mg does not interact with Sc and does not prevent the participation of the intermetallic pointed above. Mg increases the lattice parameter of the Al matrix lowering its



difference from Al_3Sc parameter to 0.7 %, which facilitates the preservation of coherency. In addition, the availability of Sc in the solid solution reduces the tendency to the precipitation of Mg_5Al_8 phase by hindering diffusion due to a strong interaction with vacancies.

In Russia there was developed a number of alloys based on the Al-Mg-Sc-Zr system with various Mg content, the alloy 1570 with 6% Mg and the additions of Mn and Be is the most strong of them. The alloy 1545 (~5.2% Mg) has found an application for welded tanks for liquefied gas.

At the IPMS and WI of NASU last time the investigations for further improving the properties of Al-Mg-Mn-Sc-Zr alloys by an additional alloying with a number of transition metals (TM) and rare-earth metals REM were undertaken. Ingots 150 mm in diameter were produced by the technique of semi-continuous casting in the PTIMA of NASU. Wrought semi-finished products and welded assemblies were manufactured in the WI.

Mechanical properties of sheets

Alloy #	Alloying elements	σ_s , MPa	σ_u , MPa	δ , %	$T_{rolling}$, °C
1	5.3Mg-0.63Mn	327	406	13.9	350
2	5.5Mg-0.70Mn-0.22Sc-0.11Zr	427	475	10.4	350
3	5.6Mg-0.62Mn-0.27Sc-0.10Zr-TM-P3M	483	513	7.8	350
3x	-"-	510	547	6.5	350 + 20

The hot-rolled sheets #1-3 after a weak cold-working were about 3 mm in thickness. The sheet #3x was produced by cold rolling of the sheet #3 to a thickness of 2.15 mm. For such sheet σ_s/ρ was 191.

High strength was obtained due to the formation in ingots of a fine-grained polyhedral structure, and in sheets of very fine ($0.15 \mu\text{m}$ in thickness) misoriented dislocation cells with high dislocation density ($to 9 \times 10^9 \text{ cm}^{-2}$) and a bending deformation in the cell body, as well as due to the precipitation during hot rolling of hardening coherent particles less than 6 nm in diameter. For the sheet #3 after a thermal treatment the strength of the welded joint of 441 MPa was achieved. In the sheet after a thermal treatment alloying by scandium lowered the tendency to the formation of hot cracks in welding process to a minimum level of 0-5% (Houldcroft test).



STRUCTURE STATE AND WEAR RESISTANCE OF HIGH-STRENGTH STEEL

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Steel products wear is the multi-factor process to be determined by normal and tangential stresses inside the layers being in contact with their interaction, by thermal phenomena occurring in them, by presence of different liquid and abrasive matters within the contact area, etc. To increase the wear resistance of steel products to be used under heavy stress conditions (e.g. railway wheels, rails, etc.) it is quite necessary to take into account the requirements of proper level of material plasticity and toughness reserve insuring brittle fracture resistance of friction surface as well as the product in whole.

So, just for such heavy operation conditions the authors have investigated the effects of chemical composition of steel, (mainly the variation in carbon content) steel structure state, hardness and toughness on wear-resistance of steel.

The wear-resistance tests have been carried-out using specific test-machine allowed to simulate the rolling friction process of two test pieces and to change the value of normal and tangential stresses of the pieces in contact. The experimental samples for these tests have been made from carbon steel material of various carbon content as heat-treated (quenching and tempering) to produce some different hardness levels and the structure including grain and lamella carbides with different dispersion. Simultaneously there have been produced some specific samples intended for definition of fracture toughness of steel with different strength and different structure state. As a parameter for this the authors used the fracture toughness k_{Ic} .

The researches performed have shown that the chemical composition of steel, its structure condition, hardness and toughness had made some significant effect on wear-resistance under the conditions of rolling-friction together with slipping phenomena occurred between the test pieces in contact. On the other hand the structure state had also made a great effect on the fracture toughness being the main parameter of reliability of the metal product under operation.

1. It has been shown that steel wear-resistance level increased with increase in carbon content and hardness of steel.
2. As a result of heat-treatment the ferrite-carbide structure with lamella carbide phase have produced making a great contribution in high level wear-resistance as to compare it with the grain-carbide structure up to 350-400 HB hardness values.
3. The increase of fracture toughness k_{Ic} increased similarly. However, even with more favorable state of structure this parameter is affected by increased carbon content in steel.



4. There have been found the optimum ratio of hardness values of two surfaces in contact, which proposed the best wear-resistance of both items equal to 1:1 with 320-360 HB as absolute values.

The relationships found out could be used, as a base for selection of structure state and properties required for obtaining high wear-resistance of items in their contact under rolling friction. However, in selection some definite technological solutions even for the conditions of the wear-resistance type under consideration one should take into account any other operational factors, e.g. dynamic and thermal stresses, operating temperature as well as the conditions required for high construction strength of metal products. The whole range of requirements shall promise high reliability reserve together with long service-life of metal products.



MANUFACTURING PARTS FOR MATERIALS-HANDLING MACHINES
WITH METHODS OF POWDER METALLURGY

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The experience of Using the methods of powder metallurgy for development of production processes for manufacturing parts of constructive designation has proved the expediency of their use in manufacturing parts for materials - handling machines. Advantages for manufacturing parts of MHM from powders are: decrease in volume of machining by cutting and material loses, increase in coefficient of material using, decrease in consumption of energy resources and prime cost of manufacturing the parts. It is possible to recommend for production parts of cylinder - cal, tapered, worm reducers; bearing caps, hatch doors, bushing spacers, worm wheel bosses, gears and rack wheels up to HB 350.

Operational designation determines the shape of a part and type of loading: static or dynamic. Requirements to strength and plasticity of the material of reducer parts are at the level of cast deformed metal. Operational properties of parts are provided with respective production process.

For manufacturing parts it is advisable to use the mixture of iron base : Жгр 1,5Д2 (Fe base; 1,2-1,8%С; 1,5-2,0%Cu); Жгр0,5ХРО,45Д3 (Fe base; 0,3-0,7%С; 1,0%Cr; 0,2-0,6%P; 1,4%Cu); Жгр0,5ХРО,45Д3 (Fe base; 0,3-0,7%С; 1,0%Cr; 0,2-0,6%P; 1,4%Cu); Жгр0,5ХРО,45Н0,8Д3 (Fe base; 0,3-0,7%С; 1,0%Cr; 0,2-0,6%P; 0,3-0,8%N; 1,4%Cu);

Materials for preparation of mixture: iron powder ПЖВ2, В3, Р3 GOST 9849-86; copper powder, grade ПМС-1 GOST 4960-75; ferrochromium, grade ФХ800 GOST 4757-89 (size category is 5,6,7); blast furnace ferrochrome ФФ16, ФФ18, ТУ 14-5-72-80, electrolytic nickel ПНЕ-1, ПНЕ-2 GOST 9722-79, pencil graphite ГК1,2,3 GOST 4404-78.

Construction parts should be manufactured according to presented production processes.

Production process №1: preparation of mixture, cold pressing powdery blanks, sintering the blanks. Production process №2: preparation of mixture, cold pressing powdery blanks, hot stamping, thermal treatment.

It is economically profitable to use technological process №1 for low loaded parts of a reducer with relative density of 70-80% and existing operating loads which do not exceed 20-25% of strength limit in cast metal.

Average and heavy loaded parts are to be manufactured by using production process №2.



Using the methods of powder metallurgy in manufacturing reducer parts for materials – handling machines will enable to decrease materials consumption and volume of machining by cutting, reduce the prime cost while producing mentioned above articles.



INTERATOMIC INTERACTIONS IN β -SiC:Ti(i)

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Silicon carbide is a promising wide-band semiconducting material possessing high heat conductivity, mechanical strength, thermal and radiation stability [1]. Introduction of 3d-metals into a silicon carbide substrate during synthesis or additional treatment makes it possible to vary the properties of the obtained materials substantially. It is known that mechanical properties of a SiC matrix can be considerably improved by introduction of TiC as the second phase [2], and vice versa some studies show that Ti–Ni coatings can be strengthened in the presence of such components as SiC, TiC₄, TiB₂.

Interatomic interactions in Si–(Me3d)–C systems were previously studied both by experimental and theoretical techniques. It was shown [3] that the interaction between SiC and Ti powders at temperatures from 1073 to 1523 (K) is very sensitive to experiment conditions. Thin SiC layers were formed at a relatively low temperature, whereas under reaction conditions TiC–Ti₅Si₃ compositions were found. SiC is characterized by low compressibility of crystals. However, hexagonal 6H-SiC crystals with traces of Ti atoms exhibit a superhigh sensitivity of ABC-bands in the photoluminescent spectrum to applied pressure [4], which is ascribed to 20% difference in the dimensions of impurity (titanium) and matrix (silicon) atoms. Si₃N₄–TiC reaction products contain mainly SiC and Ti(C,N) phases. It was found by the model pseudopotential method that for the SiC(001)/Ti interface Ti–C bonds are characterized by strong covalent C2p– and Ti3d–AO interactions [5]. In the case of Si interface Si–Ti bonds are of a metallic character [6].

In this work the presence of a neutral titanium atom impurity in an interstitial Si-T_d site was simulated by the [Si₄Ti(i)C₁₀]²⁴⁻ cluster. The X _{α} discrete variation (X _{α} -DV) method [7] was used as the main method for calculating the electronic structure parameters. The computations were performed in the context of the electronic density functional theory by the self-consistent spin-unrestriction method. Numerical integration was carried out over a set of 6000 points. The exchange-correlation potential was used in the Gunnarsson–Lundquist form. Orbital and overlap populations were analyzed in the framework of the Mulliken method. The following parameters have been calculated: 1) atomic orbital populations; 2) charges of cluster atoms; 3) magnetic moments; 4) total energy and its components; 5) total (atom – atom) and partial (AO – AO) bond populations; 6) total and partial (Me3d–AO) densities of states projected on to an energy scale.



Analysis of the data obtained allows us to make the following conclusions about the electronic structure and chemical bonding peculiarities brought about by the interstitial defect of the $Ti \rightarrow Si-T_4$ type.

The coordination polyhedron of a titanium atom in the considered position is formed by silicon atoms, which are the centers of carbon tetrahedra. Some of these atoms that bind SiC_4 tetrahedra in a chain constitute the second coordination sphere of the interstitial titanium atom. They are six in number and form an octahedron, the symmetry of the environment corresponding to O_h of the point group. According to the calculations performed, the impurity atom interacts mainly with silicon atoms. Ti-Si bond populations vary slightly from 0.117 to 0.125 e, whereas Ti-C bonds can be characterized to a greater extent as ionic, since bond populations of this type are an order of magnitude smaller. A more detailed analysis of AO - AO interactions shows that $Ti3d-Si3p-AO$ interactions are the most significant ones, whereas $Ti3d-Si3s-AO$ interactions are two times weaker, and $Ti4s-4p-AO$ interactions make a negative contribution to the formation of chemical bonds Ti-Si, $Ti4p-AO$ being more active as compared with $Ti4s-AO$. However, these interactions remain weaker than the interactions with $Ti3d-AO$.

Titanium atom in the considered interstitial site is in the non-magnetic state. The $3d$ -band of this atom is hardly split by the crystal field of neighboring atoms, it has a very small width and lies chiefly below the Fermi level. The calculated charge of titanium atom (+1.02 e) can be explained by partial admixture of titanium AO to $C2p-AO$ and by subsequent charge transfer.

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HEAT EMBRITTLEMENT AND MECHANISMS OF BOILER STEEL FAILURE

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The increase of power force equipment reliability was always the important problem, but now the special importance is attached to it. The strictness of temperature and power modes of operation, on the one hand, and the decrease of water resources, and also the increase of requirements in environment protection, on the other hand forced to use water of return cycle of chemical productions in technological process, having inevitably the small concentrations of organic mixtures even after treatment that has lead to difficultly forecasting water action of some factors which were not taken into account earlier.

In this work the influence of organic pollutants, more often contained in water chemical mixtures, on a change of structure and degradation of steel 12X1MF widely used in home power machine-building construction for the production of water piping with high pressure, heat-exchangers and etc has been investigated.

Like the operational the thermopower conditions have been modulated in watertight closed (welded) bomb from a piece of tube from steel mentioned above. Total volume of such a bomb was $8 \cdot 10^{-4}$ m³, the specimens for mechanical tests were loaded and water with admixtures of potential pollutants was poured.

The cycle consisted of heating during 8 hours up to t540° and pressure of 14 MPa and of cooling during 16 hours.

It is stated that therobaracycling brings to, on the one hand, not so large increase of temporary resistance of breakage by 2...5%, and on the other hand, to a decrease of relative elongation and to relative karrowing.

The results obtained do not contradict with modern ideas about development of heat brittleness in chromium low-alloyed steels according to which the decrease of plasticity may be caused at the same time by decay of supersaturated of hard solutions and by coagulation of carbides. These processes undoubtedly must occur and in the presence of original hydrogen-content pollutants. With working temperatures and pressures in inner surfaces of tubes the destruction of organic molecules occurs and the diffusion of hydrogen being formed in this time into the surface and into the depth of material is being taken place too. Its participation in processes of carbides coagulation accelerates the degradation of plasticity properties.

Fractographical analysis of fractures showed that after similar quantities of cycles (93 cycles) the character of steel destruction depends on the type of organic



admixtures. So, the thermocycling in technical water retains the viscosity of fracture of initial structure of steel. The surface of specimen fracture after thermocycling in water with admixtures of PAV and three – ethyleneglucole consists of facets of transcrystalline chipping and intercrystalline fracture. The extraction of carbide particles is in inside and along boundaries, they are seen in a better way after thermocycling in water with tricrezylephosphate. The particles of carbides, extracted at the boundaries of grains in the process of long stand, coagulate and are extended along the boundaries increasing the portion of intercrystalline fracture.

Therefore, the inflow of free hydrogen to the steel surface formed as a result of organic admixtures under the action of high temperature and pressure brings the changes in a state of structure elements of steel 12XIMF. Some part of free hydrogen first is desolved in ferrite, then is extracted as the particles of cementite. Further increase of stand realizes the step-by-step transformation of isomorphous matrix of carbide M_3C into special carbide $M_{23}C_6$ and the loss due to the coherent connection of carbide-matrix. Viscous character of fracture is transformed into brittle one, here the activity is defined by chemical stability of organic admixture. In real situation it means the quick and even catastrophic fracture of tubes and units of steam-pipings.



**FALURE OF THE SURFACE LAYERS OF TITANIUM ALLOYS
MODIFIED BY INTERSTITIAL ATOMS
UDER THE CONDITIONS OF FRICTION AND WEAR**

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Thermodiffusive saturation of titanium alloys by interstitial impurities (oxygen, nitrogen, carbon, boron) needs high temperatures and long-term exposures to obtain the high values of surface strengthening. The intensification of surface strengthening contributes to productivity and effectiveness increasing of its technological process.

At the moment the intensification processes of titanium alloys' thermodiffusive saturation are described due to new thermal oxidation technology [1]. Boiling bed oxidation and vacuum oxidation allow to reduce the exposure and drop off the temperature of the process in order to get certain strengthening grade and depth. The hard wear resistance diffusive layer with microhardness of 6...8.5 GPa (depends on alloys) is formed by boiling bed oxidation at 650...800 °C for 4...7 hrs. The surface layer consists of TiO_2 (rutile) oxides scales as the outer layer of 3...7 μm in thickness and interstitial oxygen solution in titanium as inner layer of 20...70 μm in thickness. During vacuum oxidation at 700...1050 °C for 0.3...0.7 hrs the surface saturation by oxygen, nitrogen, carbon results in the formation of the hard wear resistance layer consisted of complex titanium oxides, nitrides, carbides and interstitial solutions theirs in metal.

The nitrogen partial pressure decreasing (0.1...1 Pa) promotes the nitridation intensification due to reduce or elimination of nitride film as a barrier [2]. Using the vacuum nitridation (vacuum heat, vacuum anneal as thermal pretreatment) provides the intensification of thermodiffusive saturation of titanium too. The low residual gases pressure in vacuum (0.1...10 mPa) before nitrogen supply into reaction chamber makes titanium surface more active owing to surface oxide films dissociation. The treated approach of intensification allows to carry out the nitridation at 750...900 °C for 5...10 hrs. Such conditions of nitridation ensure the formation of strengthening layer consisted of $TiN+Ti_2N$ nitride film ($<1 \mu m$) and thick diffusive inner layer (100...180 μm).

The additional alloying of the nitrided surface by boron ensures the formation of the strengthening layer. The surface microhardness increases from 6...14 GPa to 18...20 GPa. In this case the complex thermal and chemical treatment (nitridation and subsequent boration in mixed powder of boron carbide and borax in ratio 5:1 under Ar atmosphere). The strengthening layer consists of the surface film of TiB_2 , TiN , Ti_2N , TiC and thick inner diffusive layer ($>100 \mu m$).



Friction and wear tests of "disk-shoe" type of OT4-1 titanium alloy after vacuum oxidation, nitridation and complex thermal and chemical treatment under the condition of boundary rolling friction in АМГ-10 fluid were carried out. The friction machine CML-2 was used. The test loads of 1 and 2 MPa were applied. The counter specimen was made of bronze (БрАЖ 9-4Л). The slip velocity was 0.6 m/s. The friction path was 10 km. Sliding couple went 200 m to grind and get the contact area of 90 % and more.

At the initial stage of friction the fragmentary transfer of the soft material (bronze) to oxidized surface is observed. The bronze species stick to the oxidized surface and form overgrowth. Later the overgrowth failure is developed. Then the new layer grows and its spallation goes again. Subsequently the bronze transfer to the oxidized surface becomes more and more intensive. The friction coefficient is high and unstable. Its value decreases from 0.215 during grinding to 0.200 after friction path of 3 km and then increases up to 0.229.

During the friction of nitrided surface and bronze the splitting off the species of nitride film surface is occurred. The hard nitride species stuck into the bronze surface and make the scratches on the nitrided surface. At first the friction coefficient increases from 0.185 to 0.215 and then decreases to 0.171. Finally it becomes stable.

After the complex treatment the species of the boration active powder are sintered to the titanium surface. At the grinding stage of friction the species splitting off is occurred and results in the greatest wear of the strengthening surface. At the next stage the fragmentary transfer of bronze species in a small amount with formation overgrowth is observed. The wear value is at order of magnitude lower than at the grinding stage and finally becomes stable. The friction coefficient increases from 0.229 at grinding to 0.185 at the stable stage.



LOAD ABILITY OF DETAILS FROM MATERIALS WITH HIGH SPECIFIC DURABILITY

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The materials with high specific durability are widely used for manufacture of different details: rings and bodies of a rolling of bearings; bodies of a rolling of transmissions friction; cog-wheels, which working surfaces are subjected to hardening etc. At account on durability of such details critical frequently accept loading, at which at the centre of contact there is a plastic deformation, as for example, at definition of supposed efforts at peak overloads of cog-wheels [1]. At the same time, in a series of cases of a detail remain efficient even at significant overflow of magnitude of this critical loading [2], and as limiting accept loading, at which the sizes of a platform of contact of details begin sharply to be deflected from calculated on the «elastic» formulas of a Hertz. In operation [3], being founded on the analytical decision [4], allowing to determine parameters elastic-plastic of contact interaction of details, the task by definition of limiting loading as for a case of primal contact of details on a line, and in a point is decided.

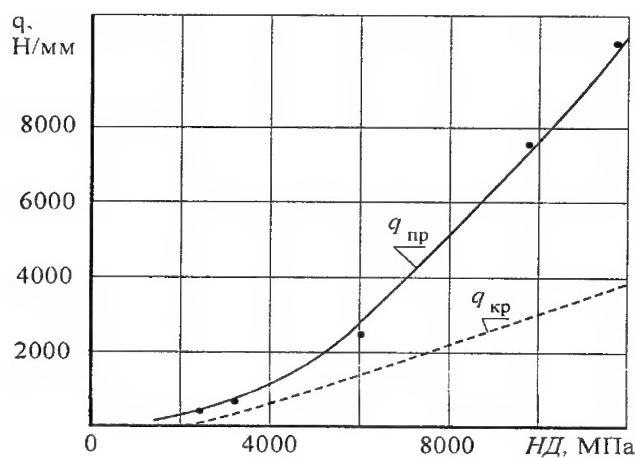


Fig. 1. Specific loading (limiting q_{np} and critical q_{kp}) in contact of a cylindrical roller ($R=5$ mm) with details of different hardness H/D ; the solid line – account q_{np} agrees [3], point – experiment; the dotted line – account q_{kp} agrees [1].



On fig.1 the graphs of dependences of specific loading q (for a case of primal contact of details on a line) from plastic hardness $H\Delta$ (ГОСТ 18835-73) materials of details are shown.

As it is visible from fig.1, at rather small hardness of contacting details of a value of specific loadings q_{kp} and q_{mp} (the origins, adequate according to the moment, of a plastic deformation on an axis of a symmetry of a platform of contact and sharp aberration of width of a platform of contact and sharp aberration of width of a platform of contact from calculated on « elastic » to the formulas) are close. At the large hardness of contacting details ($H\Delta$ 9820...11200, appropriate $HRC \geq 62 \dots 65$) values q_{mp} more, than in 2 times exceed q_{kp} . Such approach allows to reveal and to use redundancies of load ability of details with high specific durability of a material.

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**INTERRELATION BETWEEN CHARACTERISTICS OF SHORT-TERM,
LONG-TERM STATIC STRENGTHS AND CREEP OF REFRACtORY
NIOBIUM ALLOYS**

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Niobium alloys 5VMTs and 10VMTs of the system Nb-W-Mo-Zr with a solid-solution hardening feature high processibility, heat and erosion resistance, and plasticity in both deformed and recrystallized and cast states. Owing to these qualities, they are used as materials for welded structures of aerospace engineering that are operated under conditions of high temperatures and the action of aggressive gaseous media.

The present paper considers the characteristics of short-term and long-term (test time up to 10 hrs) static strengths, creep, structural state, mechanisms of plastic deformation, and the mode of fracture of sheet material, weld metal, and welded joints of niobium alloys 5VMTs (grade 5V2MTs-2) and 10VMTs and their modifications within the temperature range from 1520 to 2270 K.

Mechanical characteristics were determined from the results of tensile tests under short-term active and long-term static loading in vacuum not lower than 0.1 Pa. The structure of alloys, mode of fracture, and mechanisms of plastic deformation were studied by the methods of optical metallography and scanning and transmission electron microscopy using foils and 2-step plastic-carbon replicas.

Alloys 5VMTs and 10VMTs in as-received condition (annealing at a temperature of 1670 K for 2 hrs) are isotropic, the structure of the material is fine-grained with equiaxed polyhedral grains of size from 20 to 25 μ . The welds have cellular structure that is characteristic of the given class of materials.

The mechanism of plastic deformation and the mode of fracture of niobium alloys are shown to be governed by the test temperature, chemical composition of the material, and its structural state. Depending on the above factors, the strength of niobium alloys under conditions of short- and long-term static loading at the temperatures above 0.5 T_{melt} is controlled by two main mechanisms of plastic deformation each corresponding to its dominant mode of fracture. In the former case, plastic deformation is realized mainly due to grain-boundary slipping. Here, a viscous grain-boundary mode of fracture generally takes place. In the latter case, a uniform single and multiple intergranular slip prevails and mainly ductile failure occurs over the body of the grain.



On the basis of the analysis of the experimental data on short- and long-term static strengths and creep of niobium alloys and assuming the intensity of thermal softening for both modes of loading in a certain temperature and load region to be identical, correlation dependences were obtained of the steady-state creep rate and time to fracture on the value of the ratio between the stresses applied and the ultimate strength of the material at a corresponding temperature. For alloy 10VMTs, the generalized empirical relations among the characteristics of short-term, long-term static strengths, plasticity, and creep were established such that include the experimental data associated with both modes of fracture and plastic deformation mechanisms that govern them.

The obtained correlation relations make it possible to predict the characteristics of high-temperature strength of expensive niobium alloys basing on more readily accessible data on their short-term strength and plasticity.



**STRUCTURE AND PROPERTIES OF EQUIATOMIC TITANIUM
ALUMINIDE, PRODUSED BY THE METHOD OF ELECTROSLAG
REMELTING**

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In present work authors fulfilled the investigations of possibility of intermetallic alloy on γ -TiAl base producing by the method of electroslag remelting (ESR) under protective atmosphere. The developed technology of equiatomic titanium aluminide (50% at. Al and 50% at. Ti) includes remelting of compound electrode from titanium VT-1 grade and aluminium A5 grade in chamber type ESR unit with using of active metal-containing flux and protective atmosphere. Obtained ingot has following size: length – 200 mm, diameter – 95 mm. Ingot was sheared along vertical axis, after that the cross-sectional templates were made. The microstructure of ingot was investigated too. The samples were cutted from cross-sectional template of the middle part of ingot. They were subjected to mechanical grinding and etching in solution of HF+HNO₃ to detect the microstructure and fulfill X-ray structure analysis. The density of material was determined by the method of hydrostatic weighing. Evaluation of corrosion resistance of material to high-temperature gas oxidation was fulfilled by the weight method (by the increasing of sample mass measurement). To determine the phase composition, the X-ray phase analysis on unit DRON-3 was carried out. It was fulfilled with using of Bragg-Brentano method in angle interval $2\theta=40\text{--}140$. Filtered Fe-beaming was used. The registration of signal was made by point method with computer recording. The step of detector moving was $0,05 \text{ и } 0,01^\circ$ for narrow and wide intervals correspondingly. Time of exposition was 5 sec. From experimental results the K α_2 component of beaming was subtracted by the computer transformation procedure. Taking into account existence of skin and core zones in macrostructure of ingot, samples for investigations were taken from surface, middle and central zones of cross-section.

After that the investigation of microstructure was carried out with using of optical microscope NEOPHOT-32 and digital camera CASIO QV-100. Digitized images were analyzed with using of image analysis software IMAGE TOOL to obtain the quantitative characteristics of structure. The measurement of microhardness was fulfilled on PMT-3 devise, 30 measurements per every point was made with following statistical evaluation. Error of measurement was less than 5%.

Because of obtained ingot has high porosity and brittleness, the attempt to investigate the influence of high-pressure treatment on structure and properties of



titanium intermetallide was made. The peculiarities of such materials behavior during temperature deformation influence in current time are studied insufficiently, but existing information of preliminary investigations indicates about possibility of efficient increasing of mechanical characteristics in result of such treatments.

Material of ingot was milled in ball mill to obtain the particle size 3-7 microns. after that it was subjected to high-temperature sintering at pressure 100-500 kgs/sq.cm. in vacuum 10^{-3} Hg.mm. with following inspection of density, microhardness and microstructure.

From the powder the plate samples 3x10x40 mm were prepared by pressing and they were subjected to high temperature sintering at temperatures 1260 (regime I) и 1420 °C (regime II), and two-stage regime (2 subsequent exposures at 1260 and 1420 °C with intermediate cooling, regime III). In all cases after sintering samples were cooled slowly with furnace with rate 80-100 °C/hour. The samples, sintered at temperature 1420 °C. It permits to say about exceeding of solidus temperature for investigated alloy, but accordingly to equilibrium diagramme this temperature must be 1450°C approximately. This fact may be explained by the influence of high pressure during sintering, that causing the decreasing of phase transformation temperature.

The fulfilled work has shown the principal possibility of intermetallic alloy on equiatomic composition base producing by the method of electroslag remelting (ESR) under protective atmosphere with using of active flux.

1. Obtained ingot had typical cast structure with high porosity (up to 15%). Ingot had high chemical and structure uniformity. It was confirmed by investigations of microstructure, X-ray analysys and measurement of microhardness.
2. It was confirmed the effectiveness of using of high-temperature sintering at high pressure for treatment of such low-plasticity materials as titanium aluminides. The proper regime of sintering permits to form fine-grained structure with low porosity.



**INFLUENCE OF ALLOYING BY SCANDIUM AND ZIRCONIUM
ON THE STRUCTURE AND MECHANICAL PROPERTIES
OF CAST AND POWDER Al-Zn-Mg-Cu ALLOY**

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Alloying cast wrought aluminum alloys by scandium is known to lead to a significant improving of mechanical properties caused by the creation of a fine-grained polyhedral structure in the ingot as well as by a precipitation from the supersaturated solid solution of disperse coherent Al_3Sc particles, which provide disperse hardening and delay of recrystallization. These effects are increased by the availability in the alloy of a small addition of zirconium. The creation of a fine-grained structure and the delay of recrystallization can be provided also by producing the alloy by powder metallurgy technique. On the other hand, by producing powders by atomization from the melt it is possible to increase the concentration of the supersaturated solid solution due to high crystallization rate, which should result in increasing a hardening effect during aging.

In the work there is undertaken a comparative study of changing structure and mechanical properties of extruded rods from a high-strength alloy Al-9.5Zn-3.0Mg-1.2Cu (wt. %), fabricated from cast and powder billets, under the influence of alloying by Zr and Sc.

Ingots 25 mm in diameter were produced by casting to thick-wall water-cooled copper moulds, which provides a cooling rate during crystallization v_r to 10^2 °C/s. The atomization of the melt for producing powders was carried out by high-pressure water, which permits to increase v_r to 10^6 °C/s. From powders a fraction of (-100+63) μm was extracted. The dried powders were pressed in the air at room temperature, and briquettes with the porosity about 30% were obtained. The briquettes were placed into a die that was introduced into the working chamber of the press, where vacuum degassing and forging (pulse pressing) with a developed pressure to 1 GPa were carried out. Before forging a briquette was hold at a temperature of 450°C during 1 h. There were produced billets 25 mm in diameter and about 30 mm in height with the porosity of 0.4-0.7 % that testifies to a failure of oxide films in powder surface in the process of vacuum forging.

Ingots and billets were extruded to a rod 6 mm in diameter at temperatures: ingots - of 400°C, powder billets - of 450°C with holding of 1 h at the extrusion temperature. Specimens for tensile test with gauge diameter of 3



mm and gauge length of 15 mm were subjected to the thermal treatment of T6 type: quenching into the water from 465°C with the following aging at 120°C during 24 h.

Mechanical properties of rods from cast (C) and powder (P)
Al-9,5Zn-3,0Mg-1,2Cu alloys after T6 treatment

Alloy #	Additional alloying (wt. %)	σ_u , MPa	$\sigma_{0,2}$, MPa	δ , %	HV, MPa
C1	-	554	-	0	1920
C2	0.2Zr	710	691	5.0	2020
C3	0.2Zr+0.3Sc	790	760	10.0	2010
P1	-	763	742	4.5	2150
P2	0.2Zr	762	749	3.7	2060
P3	0.2Zr+0.3Sc	779	751	5.5	2100

A brittle fracture of the C1 rod is caused by its recrystallization while heating for quenching, the fracture occurred mainly in grain boundaries. In P1 rod only separate recrystallized areas were seen, the structure was mainly a cellular polygonized one that provided the possibility of plastic deformation while tensile test. The best combination of high strength with a comparatively high ductility is observed in the rod of the C3 alloy. Here the alloying by Sc+Zr transformed ingot structure from a dendrite one (C1 and C2 ingots) to a polyhedral one with the grain size about 25 μm , and in the rod after T6 treatment it completely suppressed the recrystallization and formed the most distinct and uniform dislocation cell structure (cell cross size about 1.5 μm) with the smallest second phase particles.

Producing powders with high cooling rate led to a sharp refinement of the dendrite structure: in various powder particles the size of dendrite branches was from 0.5-2 μm to 2-5 μm contrary to 100-200 μm and 20-50 μm in the ingots C1 and C2 respectively. In powder particles of P3 alloy due to high v, primary Al_3Sc particles that are nuclei of crystallization in the case of the formation of a polyhedral structure were not formed, the structure of P3 particles remained dendrite with the increase of the amount of particles with small dendrite branches. The above-mentioned factors as well as the availability in the rods from powder alloys of a large amount of former interparticle boundaries led to the creation in them of less uniform cell structure and less uniform precipitates (probably including oxide phases), which has somewhat lowered the plasticity of powder alloys and increased strength in powder alloys without Sc.



EFFECTIVE GRAIN SIZE IN DEFORMED METALS

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In the applied relation the considerable interest represents a study of a combination of grain boundary and deformation strengthening (work hardening) mechanisms in metals and alloys. The idea of work is formulated simply: how at formation of deformation structure to receive the minimum possible size of cells, while its boundaries to make to work as grain boundaries, as the grain boundary strengthening is most effective.

It is known, that the dislocation structure during plastic deformation of metals transits, as demonstrates electron-microscopic researches, through series stages of "wood" dislocations, formation of tangles and formation of "cellular" structure, which in turn can be light disorienting and strongly disorienting at more late stage of deformation (so-called, recovery structure). In work it is shown, that the boundaries of cells only then start to play a role of grain boundaries, when the recovery structures and polygonal cell boundaries are formed.

There is a known experimental fact, that the diagrams of a strain for all sizes of grains (rebuilt in coordinates true stress S - true strain e) coincide at a constant temperature, starting with III stage of hardening. The diagrams coincide because represent the same curve of the second order (parabola) with an index $n=0,5$ (for BCC - metals), or $0,5 < n < 1$ (for other types of crystalline lattices). Besides they have identical coefficient of deformation hardening K_3 in the equation

$$\sigma = \sigma_0 + K_3 \sqrt[n]{e} .$$

It allows to combine at superimposition (!) the stress-strain curves of recrystallized and predeformed samples and to combine (with precision 1 -2 %) in coordinates $S-e$ only on a unique part. Such overlapping of the diagrams obtained at low temperature (usual, at a room temperature or at 100 °C for metals with a restricted plasticity), enables to determine not only effective grain size of the given structure, but also equivalent strain, i.e. strain, which on a level of a reached dislocation density is equivalent to all before carried out thermomechanical treatment. The overlapping of the stress-strain curves is possible and for samples with the different grain sizes, as coefficient K_3 does not depend on a grain size if to view a part of the diagram, bounded only with strengthening. However on amplitude the distinction of stresses here can be very major because of Petch dependence of a yield strength σ_s from a grain size.

Two molybdenum alloys (CM10 and CM12) and armco-iron were researched with the purpose to receive of the minimum possible effective grain sizes.



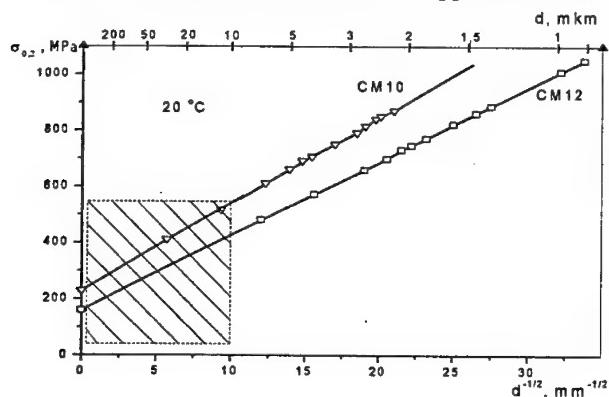
On a molybdenum alloys tested at temperatures 20 and 100 °C, were constructed, based on overlapping technique, offered in work, Petch of dependence with coefficients $K_y = 18.5 \text{ MPa/mm}^{3/2}$. For an interval of the grain sizes from 1 mm and up to 15 microns the experimental data on definition of a yield strength will be utilized at the different grain sizes, while the technique, offered in work, is applied for more small-sized grains.

The first and main result of the work consists that obtained dependence $\sigma - d^{-1/2}$ (look a figure) uncloses extensive opportunities for practice. It allows in 2 - 3

times to increase a level of useful stresses (i.e. in case of molybdenum to exceed stress 1000 MPa), at which the dilute alloys can be utilized.

In addition effective grain size is reduced up to 1 micron and below, that practically it is impossible to receive by a conventional me-thod using strain and recrystallized annealing, as nucleuses of recrystallization have the greater size. For obtaining of the so small-sized effective grain sizes, the strains $\epsilon > 1$ utilized at temperatures 700-1000 °C. The similar results were obtained and for an armco-iron.

Besides in work the explanation of a limiting refinement of deformed structure and, accordingly, of the greatest possible strengthening of metals is offered and the analytical expression for exposition of this appearance is obtained. The reason is the appearance at high temperatures, which are necessary for lowering a stress level at major strain degrees and for development of processes of dynamic recovery, nucleuses of dynamic recrystallization.



KINETICS OF SOLID SOLUTION DECOMPOSITION IN Al - Mg - Sc SYSTEM

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Alloys on the base of a system Al - Mg with Mg content up to 6 wt. % are wrought well-welded alloys. These alloys have a high plasticity and corrosion resistance, but their strength is rather low. Sc is one of the most perspective components, which allows to increase strength and thermal stability of alloys and corrosion resistance as well. The efficiency of scandium increases in the presence of Zr.

The kinetics of precipitation of particles of the intermetallic compound $\text{Al}_3(\text{Sc}_{1-x}\text{Zr}_x)$ from the solid solution of alloys on the base Al - 5%Mg, as well as the formation of structure and mechanical properties depending on the chemical composition of alloys and the technology of their manufacturing are investigated in this work.

Alloys of the composition Al - 5.0Mg - xSc, where $x = (0 - 0.5)$ wt. %, and Al - 5.0Mg - 0.21Sc - yZr, where $y = (0 - 0.6)$ wt. %, were investigated. The ingots of alloys Ø 55 mm were produced by casting to a water-cooled copper mold. The cooling rate was ~ 100 K/c, that allowed to fix in the solid solution up to 0.3 % Sc. The obtained alloys were subjected to two-stage extrusion at temperatures of 573 K, 623 K, 673 K with a drawing coefficient of 4.8 and 17.4. The mechanical properties of extruded rods were investigated by measuring hardness (HV) and carrying out tensile tests on five-fold samples with gauge diameter of 3 mm. The structure was studied by TEM. Among alloys Al - 5.0Mg - xSc the maximum strength is obtained in the alloy Al - 5.0Mg - 0.3Sc ($\sigma_{0.2} = 370$ MPa, $\sigma_u = 465$ MPa at $\delta = 11.5\%$) after extrusion at 573 K. The increase of extrusion temperature, increasing the size of cells and secondary Al_3Sc particles, leads to some decrease of strength ($\sigma_{0.2} = 280$ MPa, $\sigma_u = 410$ MPa, $\delta = 16\%$ after extrusion at 723 K).

The influence of additional alloying by Zr on strength characteristics of the alloy Al - 5.0Mg - 0.21Sc was studied for rods obtained by extrusion at 623 K. The addition of 0.12 wt. % Zr has increased alloy strength by 14 %.

The change of hardness after annealing of extruded rods was investigated at 623 K, 673 K, 723 K. The softening of rods in alloys with Sc and Zr occurs more slowly, than in alloys without Zr: thus, in the alloy Al - 5.0Mg - 0.3Sc the falling-off of HV was watched after 3 hours of annealing at 623K, and in the alloy Al - 5.0Mg - 0.21Sc - 0.60Zr after 8 hours of annealing.



The research of structure has shown that the alloy Al - 5.0Mg after extrusion at 623 K and 673 K was completely recrystallized. Only after extrusion at 573 K in this alloy a fine dislocation cellular structure has formed. In alloys with Sc, the cellular structure was formed at all modes of extrusion. With the increase of extrusion temperature the cells size (D) grow. After extrusion at 723 K the size of some cells in the alloy with 0.3% Sc exceeded 10 μm , but metallographic analysis revealed no signs of recrystallization both in extruded, and annealed specimens. The additional alloying by Zr decreases the cell size. So, in alloys Al - 5.5Mg - 0.65Mn - 0.22Sc - xZr extruded at 623 K D = 1.2 μm at x = 0 and D = 0.7 μm at x = 0.6% Zr.

Growth of secondary Al_3Sc particles during annealing was studied on the example of the alloy Al - 5.0Mg - 0.21Sc - 0.12Zr (fig. 1). In the extruded rod Al_3Sc particles were revealed extremely seldom, since because of decreasing dimensional misfit of crystal lattices of the matrix solid solution and $\text{Al}_3(\text{Sc}, \text{Zr})$ in particles Al - Mg alloys the contrast on coherent particles vanishes when their diameter is lower than 5 nm.

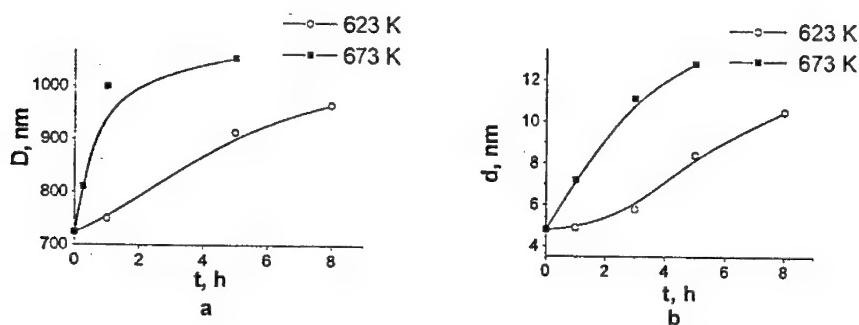


Fig. 1 - Change of structural characteristics of rods Al - 5.0Mg - 0.21Sc - 0.12Zr (extrusion temperature is 623 K) after annealings at 623 K and 673 K: a - size of cells, b - diameter of Al_3Sc particles.

Thus, the alloying by 0.12% Zr allows to reduce the content of Sc to 0.2 % and simultaneously to raise the level of mechanical properties and of thermal stability. While annealing the size of Al_3Sc particles a little increases, but they remain small-sized enough to hinder recrystallization. Due to this fact the investigated alloys preserve a high level of strength properties after continuous annealing at 623 - 723 K. The ratio d/D nevertheless essentially increases after continuous annealing at 673 K (8h), therefore it is possible to admit that more long-duration annealing can lead to a transition from slow growth of cells to their fast growth and recrystallization.



GRADIENT AND LAYERED MATERIALS

129-158



GREDIENT STRUCTURE-PHASE STATES IN RAIL STEEL AFTER MAGNITORLASM TREATMENT

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The purpose of this research was the layer-by-layer study of phase composition and defect substructure, being formed in plate pearlite steel 70XCA having the structure undergone by magnetoplasma treatment made by plasmotron equipped by system allowing to do the replacement of jet on a head of rail with frequency of 300Hz. Rail was being replaced with a speed of 1,5-2 m/s, that gave the possibility to obtain the unbroken strengthening of rolling surface to the depth of 0,8-1,2 mm.

It is stated that poor-pickling surface layer has the gradient grain structure: the average sizes of grains and also the value of their dispersion quickly decrease in remoting from treatment surface achieving its minimum value at the boundary of division of the poor-pickling layer with basic volume of steel.

In layer of 0,4-0,6 mm thickness (including and its surface) on boundaries of grains there are the interlayers of carbide phase, in inside of grains – the crystals of martensite structure. Remoting from treatment surface to the depth approximately 2 mm is accompanied by step-wise increase of average size of grain; at a distance approximately 5 mm from the surface of treatment the average size corresponds to the size of grains of initial material.

During researches carried out by methods of diffractional electronic microscopy, it was stated that the material of presurface layer of 0,4-0,5 mm thickness having the surface of treatment consists of martensite exclusively.

With remoting from treatment surface by 0,6-0,8 mm the products of quasitempering of the quenched state begin to prevail in steel structure; the crystals of martensite are broken into fragments; the particles of iron carbide, located in crystals of martensite acquire the round form and are grouped on boundaries of fragments or martensite crystals; in some cases there are areas in which one can see the recrystallization nucleation.

At a distance of approximately 1mm from treatment surface side by side with martensite one the structure is displayed being like globular pearlite by morphological feature.

It is evidently, that the phases described above: martensite, microglobular pearlite, "pseudoperlite" and plate pearlite contain the quantity of carbide phase.

The less quantity of carbide phase is in martensite structure. It is generally so-called cementite of "selftempering".

The structure-phase transformations occurring in steel 70XGSA during magnetoplasma treatment revealed by method of diffractional electronic



microscopy possess the different range of defects presence of α - phase crystalline structure.

As it was to be expected, the presurface layer having a martensite structure is the most defective. Remoting into the depth of material is accompanied by essential decrease of scalar density of dislocations and amplitude of curvature-torsion of α -phase crystalline lattice.

The quantitative researches of structure and phase composition carried out in this work allowed to estimate the steel strength 70XGSA on the range of fluidity depending on a distance to treatment surface. The range of steel fluidity in additive approximation can be studied as the sum of the following components:

$$\sigma_{0,2} = \Delta\sigma_0 + \Delta\sigma_{h...s} + \Delta\sigma_p + \Delta\sigma_{d,p} + \Delta\sigma_d + \Delta\sigma_{m,p}(\Delta\sigma_{m,pl}) + \Delta\sigma_{f,s}$$

here $\Delta\sigma_0$ is the stress of friction of α -iron lattice; $\Delta\sigma_{h...s}$ is the hardening of solid solution on the basis of ferrite by atoms of alloyed elements; $\Delta\sigma_p$ is the strengthening on account of pearlite of steel constituent structure; $\Delta\sigma_{d,p}$ is the hardening of dispersed particles of carbide phase; $\Delta\sigma_d$ is the hardening of "forest" dislocations; $+ \Delta\sigma_{m,p}(\Delta\sigma_{m,pl})$ is the hardening on account of martensite packages (martensite plates); $\Delta\sigma_{f,s}$ is the hardening of far-acting fields of stresses.

The study material is the "natural composite" having comparatively strong presurface layer, mild transitive area and core on a size of yield point occupying the intermediate position.



**SPECIAL FEATURES OF STRUCTURAL CHANGES, INDUCED BY THE
HIGH SPEED TORRENT OF PARTICLES IN BOTH IRON-NICKEL AND
IRON-MANGANESE ALLOYS**

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Metallography, X-ray diffraction, local X-ray spectrum and magnetic analysis were used to study structural transformations in both iron-nickel and iron-manganese alloys that were subjected to impulsive loading by the high speed (about 1 km/s) torrent of particles (HSPT) SiC and Cr [1]. It was established, that structure of alloys changes under influence two factors of HSPT: 1) under influence shock waves induced HSPT, in the specimen volume and 2) under effect penetrating particles, in channels and near of them. Effect of uniform deformation (twinning, formation of cell dislocation structure) and effects of local deformation caused shock waves were established. Essential heterogeneous of this types of deformation was showed. The background value pressure, equal 8 GPa, was determined from observation of alpha-gamma transformation in high nickel alloys with known value of the phase transformation critical pressure. The observation of traces of alpha-epsilon-alpha transformations in low nickel alloy indicated much more value of background pressure, 12 GPa [2]. Preliminary data testified that pressure more than 100 GPa and residual temperature about 700°C had realised in the superdeep penetration channels. As a result of the investigations of the Fe-Mn alloys, containing 10, 21 and 30% Mn, which are characterized by both the various type of crystal lattice (BCC, HCP, FCC) and the various metastability degree toward gamma-epsilon and epsilon-alpha martensitic transformations, was found that a type of crystal lattice didn't influence upon the superdeep penetration channel density in the bulk of specimens of alloys. A hardness of the materials has a visible effect on the superdeep penetration channel density in the materials. An increase of hardness of the alloys leads to a decrease of the superdeep penetration channel density. It was found that a shock wave induced in the alloys both gamma-epsilon, epsilon-alpha martensitic transformations and the noticeable reduction of the crystal lattice parameters of gamma and alpha phases. An existence of the macroscopic residual stress fields was discovered in the loaded specimens of the



Fe-Mn alloys. A complicated structure of this fields was revealed by the specific electrolytic etching of the specimens.

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MATHEMATICAL MODELLING OF A KINETICS OF HETEROPHASE ALLOYS STRAIN HARDENING AT DIFFERENT TEMPERATURES

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Large place among high-strength materials take heterophase alloys, which hardened by dispersed particles of a undeformed phase. Strength and plasticity of heterophase alloys are very sensing characteristics to a structural condition and structural transformations. The interaction of dislocations with particles apart from hardening effect results in appearance of new components of dislocation structure. The interconversions between these structure components are possible during the deformation and subsequent relaxation processes. Thus nature and the outcome of interactions of dislocation structure components with particles can vary with change of the scale characteristics ratio for a hardening phase (sizes, form of particles, spacing interval between particles) and the dislocation structure (length of a source, spacing interval between dislocations).

The physical model of dislocation structure evolution is advanced on the basis of the deformed strengthening and dynamic return concept. Two basic mechanisms of shear-formation are considered during simulation: a) the shear-formation dislocation loop, extending in process of deleting from a source, is hardly bound to slip plane - planar model; b) the shear-formation dislocation loop crosses the forest dislocations, goes out the slip plane - spatial model. The physical criterion of transition from planar to the spatial model is reviewed.

The comparison of the scale characteristics of dislocation structure and hardening has allowed to section range of a dislocation density on some intervals. 1) Area of a homogeneous deformation, when the slip dislocation is predominantly supported by external stress and is particulate reversible. 2) Areas of dislocations slip localization. There the slip zones form very quickly and residual tracks of slip appear. The expansion of localized and uniform deformation stages is determined by the sizes of a hardening phase and degree of its disperse.

Influence of stresses reverse fields on existence and expansion of an inhomogeneous deformation intervals is taken into account. Is shown, that the number of dislocations emitted by a source, is not monotonous varies on all stress-strain curve. Influence of the particles sizes and spacing intervals between them on a level of localization and on an expansion of the homogeneous and localized deformation stages is detected. The number of dislocations decreases in a slip zone at the alloys with a coarse particles.

The system of differential equations of deformation defects accumulation is recorded in different temperature intervals: 1) at low temperatures, when the diffusive relaxation processes is realized by only interstitial atoms; 2) at temperatures of bivacancy relaxation of dislocation structure; 3) at high



temperatures, when the mobility of deformation monovacancies becomes enough high and they introduce the noticeable contribution to diffusive relaxation processes. The given system of differential equations is resolved by the numerical methods in a broad interval of varied parameters.

The evolution of dislocation structure components (shear formation dislocations, prismatic interstitial and vacancy dislocations, interstitial and vacancy dipole configurations) during deformation are analyzed in the intervals of the homogeneous and localized deformation.

The influence of deform temperature on behavior of a dislocation structure different components is investigated. At low temperatures the main contribution to accumulation of a common dislocation density and, therefore, in strengthening of material is introduced by interstitial prismatic loops. And only in materials with very large particles density of the dipole configurations becomes comparable with density interstitial prismatic loops during deformation. At increase of temperature, when become mobile bivacancy, the density of interstitial prismatic loops and vacancy prismatic loops are approximately equal during deformation. In alloys with large particles the density of dipoles considerably increases during plastic deformation. Thus, the dipole configurations introduce a substantial contribution to material hardening both at low temperatures deforming and at high.

The curves of flow, conforming to varied dislocation structure are obtained. They have different behavior at various stages of deformation. The simulation by stages of dislocation structure evolution forecasts change of the strain hardening factor during deformation in actual heterophaze alloys. The estimation of resistance to dislocations motion in a slip zone is conducted, that shows satisfactory agreement with the experimental stress-strain curves.



PECULIARITIES OF STRUCTURE-PHASE TRANSFORMATIONS IN STEELS DURING LARGE PLASTIC DEFORMATIONS

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Solving the problem of controlling the strength and plasticity of solids and development of new technology process of drawing and cold volumetric stamping on this basis are connected to some extent with understanding of physical nature of deforming.

The purpose of this work is the study of conformity to structure – phase transformations during intensive plastic deformation in time of cold volumetric stamping of steel of different structural classes.

The optic and electronic microscopy was used as the research method of low-carbon steel and low-alloyed steels.

It is stated that the stamping of billet is accompanied by composite transformation of grain structure of central zone of material. In part of specimen, undergone only the extrusion, the grains are extended along the axis being parallel to it, in upper part of specimen they are compressed in longitudinal direction with simultaneous stretching in cross-section direction that changes the angular texture of materials. It should be noted that here the continuity of material in volume is retained and only by periphery of head there is the formation of quasiperiodicity placed by longitudinal cracks.

Analysis of pictures of structural texture originated after each technology operation of bolt production gives the untensive picture of distribution of plastic deformation and therefore the distribution of stresses in different places of forming product. Although the bolt is too simple detail and during its production the simple types of deforming are used, the investigation shows not very trivial picture of deformation localization on macrolevel and on the level of grains groups in different stages of processing. It should be necessary to note main features of decribed phenomenon. First, inhomoginity of local deformation is very impotent. By absolute value the deformation amplitude changes from 0 to 300% and besides its sign can be different ("plus" ore "minus") on various technology operations. Second, on a large part the maximum deformation is concentrated in surface areas of products with quick decrease of it into body of billet. Third, the vector of structural texture takes different directions from 0 to 160 degrees in various stages of product formation. Fourth, the gradient of vector orientation of structural texture is too large in the vicinity of areas of deformation localization. Firth, in places of local gradient of deformation rate and vector orientation of structural texture the cracks are uncovered.



Then the dislocation structure being formed in different stages of producing the composite detail with large gradients of plastic deformation has been studied quantitatively for the first time. It is stated that the large number of substructure parameters depends only on the rate of plastic deformation. The fragmental structure occupies nearly 100% of material in fully formed product. Inside of fragments the cell-lattice structure begin to be formed.

The process of cementite fracture during plastic deformation occurs in some stages on different structural levels. It consists of fracture of pearlite grains, crushing of pearlite colonies, cutting and grinding of cementite plates upto formation of nanomeasuring particles. The structural levels from grains to microlevel are embraced by this process. It is shown that the mechanism of fracture of the pearlite colonies depends on their orientation in relation to axis of deformation. Different stability to plastic deformation of cementite in pearlite colonies and in interlayers on boundaries is stated and it is noted that the particles on boundaries of grains are much stable (Table 1).

Structure-phase transformations occurring in steels 20G2P in process of intensive plastic deformation

Substructure type	Pearlite state	Cementite on boundaries of grains
Chaos of dislocation	Pearlite grains drawn along pearlite grains	Perfect elongated Particles of cementite (interlayers)
Cell-net. Anisotropic fragments with cell-net substructure large amplitude of curvature-torsion of crystalline lattice and phase.	"Pseudoperlite" pearlite is nearly disintegrated.	Interlayers "are defected" fracture of interlayers is finished. Round particles begin to appear.
Well developed fragments	Subboundaries in places of cementite pearlite.	Large undefragmented round particles of cementite with high level of local fields of stresses.



**STUDY OF INFLUENCE OF A DIFFERENTIAL QUENCHING
ON STRUCTURE AND PROPERTIES OF RAIL
STEEL FROM 0,71 % C AND 0,75 % Mn**

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One of perspective methods of a heat treatment become is the differential quenching, at which the increase of durability is reached as the sum of effects of local and general influence on metal. Now already it is obvious, that the uniformity of metal materials not always is desirable. The large practical interest is represented by gradient structures in surface layers created by a differential quenching. All complex of properties of a material, in which themes or different way the gradient structure (parameters is created and which properties change with depth), in most cases answers to real conditions of operation, than property of a homogeneous material much better. The differential quenching forms in a surface layer the structure having raised hardness and durability, and softer structure in other material. Cooling at a differential quenching usually carry out by a mix of water with air or flows of water - at first intensive cooling of surface layers up to temperature 400 - 500 °C, then rather slow cooling of all volume of metal.

The purpose of work was research of gradient structures in rail steel, after a differential quenching and study of influence of such quenching on properties of steel. Samples of steel from 0,71 % C and 0,75 % Mn, prepared from rails were exposed to a heat treatment. The quenching in two cooling environments was carried out from temperature 840-860 °C. Temperature of a sample in at heating and during cooling was supervised on different depth from a surface, which cooled. After endurance for an austenization the head of a sample up to middle of its lateral side placed on first cooling environment (water, water solution of salt etc.), then, at achievement by surface layers of metal on depth of 5-10 mm from a surface of temperature ~ 500 °C, a sample completely placed on second cooling environment (oil). Structure and structure of samples were investigated on thin samples by standard optical and electronic methods.

Is shown, that the differential quenching of rail steel from 0,71 % C and 0,75 % Mn forms in a sample of a rail a rather thin surface layer (depth of 1-2 mm) raised hardness (on 3-5 HRC above, than at a standard quenching; the hardness smoothly decreases from a surface and lateral sides of a sample of a rail to his middle), the durabilities and resistance (in 3 times are higher usual) with a well appreciable gradient structure consisting of mixes of ferrite and a cementite with various distance between plates of ferrite and a cementite (structure of a troostite - at a surface, lamellar pearlite - on depth > 2 mm), free ferrite and cementite located on borders of grains. With increase of distance up to a surface the relative quantity of



a pearlite grows, reaching 100 % on depth of 3 mm; relative quantity of free ferrite and cementite on borders of grains are reduced. Within the limits of a surface layer by thickness 1,0 - 1,5 mm are revealed characteristic unmonotonous dependences of the size of a grain and distance between separate plates of ferrite and cementite from distance up to a surface, which subjected to processing. On a surface become the structure consisting of a pearlite with the size of grains 0,0043 mm is formed; on depth of 0,2 mm the size of a grain reaches 0,0057 mm; further (up to 0,65 mm) the grain again decreases up to the size on a surface, then grows, reaching constant meaning on distance of 5 mm from a surface. The structure of a pearlite has greatest distance between plates of ferrite and cementite on a surface (up to depth 0,13 - 0,15 mm this distance ~ of 0,0001 mm); then distance between plates of ferrite and cementite is increased up to 0,00014 mm and further after insignificant reduction gradually grows, up to meanings, characteristic for a lamellar pearlite of the basic material. The heterogeneity of density of defects of superficial and deeper layers of steel after a differential quenching is marked. Greatest density of defects is present in a surface layer; the removal deep into of material is accompanied by essential decrease of density of defects. Thus and surface layer created by a differential quenching is not homogeneous on structure and on properties. The unmonotonous character of the received dependences, obviously, is connected to a tempering of surface layers after a quenching and processes similar to a recrystallization at heating of metal after processing. The received data and results of mathematical modeling on the COMPUTER testify to correctness of this conclusion and volume, that the formation of gradient structures is explained by special temperature conditions of processes of disintegration of an austenite at a differential quenching of rail steel. The rail steel subjected a differential quenching, represents the special non-uniform material having rather firm and strong surface layer, rather soft transitive site and middle with intermediate meanings of durability and hardness.



**PECULIARITIES OF STRUCTURE AND PROPERTIES OF THE
FIBROUS TITANIUM-ALUMINIDE COMPOSITES, PRODUCED BY
THE METHOD OF PACKET PRESSING**

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In last years the interest to intermetallides as structural materials has increased essentially. But their low plasticity at room temperature makes difficult to obtain the part with predetermined shape and restrains the commercial application of the such materials [1]. In paper [2] the technology of producing of the parts from titanium aluminide was proposed. It includes the manufacturing of ingot by the electroslag remelting process, milling in ball mill and following hot pressing and sintering.

We have studied the possibility of intermetallic parts producing from the blanks of pseudo-alloy, that were obtained by the common hydroextrusion of wire titanium and aluminium elements, assembled in packet according to hexagonal scheme. Due to activation of intercomponent diffusion during the common plastic deformation with elongations 10^4 - 10^5 at contact stresses 1-2,5 Gpa the complex of mechanical properties, necessary for further change of shape was obtained. For following transformation of pseudo-alloy to intermetallide (with possible inheritance of fibrous structure) the annealing at temperature of peritectic transformation was carried out. The additional advantage of packet scheme is absence of strict demands to atmosphere of furnace during phase transformation because the density of preformed pseudo-alloy is close to theoretical determined by the components volume ratio and surface of pressed blank is defended by thin-wall case, that may be removed after discharge from the furnace.

With using of 3-stage regime of pseudu-alloy producing, the diameter of filaments bring to 1-15 microns. Softening tempering (420 - 450°C) had fulfilled at first stage only in order to prevent the intermetallide formation.

The pseudo-alloy bulks were investigated by the methods of optical microscopy, X-ray spectrum analysis and measurement of microhardness.

X-ray spectrum analysis shows that at first stage the zone of component interaction 3-4 microns wide forms and its width practically does not increase in result of following deformation. Apparently, ratio of zone of element interaction width to their cross section size is the main parameter, that determines the success of further changes of shape.

The microhardness of different structure components increases in 1,5-1,7 times at first stage of deformation, does not change at second stage and decreases nearly to initial level at the end of third stage. It may be explained by the



development of processes of dynamic recrystallization. The formation of intermetallides at this stage have not been found, but the view of elements distribution curve does not exclude their existence. This fact is not desirable for further formation of part shape.

During free upset, the degree of deformation at the destroying of the pseudo-alloy bulk was 35%. At bulk hydrostatic stamping (0,8 Gpa) this characteristic was 73%. It is enough for rather wide range of products.

The kinetics of intermetallide formation was studied at temperatures 700, 800, 900, 1000°C during exposure from 25 minutes to 4 hours. Treatment at temperature 700°C during 1 hour leads to decreasing of titanium grains hardness for 50-60% when hardness in boundary zone and in aluminium component keeps at the initial level. After 2 times increasing of exposure at this temperature, hardness of boundary zone increases to 3,3-3,5 Gpa and noticeable amount of intermetallide (15-20%) was fixed after exposure 3 hours. The needles of intermetallide were 200-250 microns in length and they have central bulge 20-30 microns (hardness 5,7 – 5,8 Gpa). The surrounding grains have size from 6 to 70 microns. After following increasing of exposure up to 4 hours at 700°C the size of needles decreases and brittleness increases.

The short heatings (25 min) at temperatures 800-1000°C form the granular structure (6-15 microns) and dendrite-like structures, formed by fine precipitation of particles with rather low hardness (0,6 Gpa). Exposure at 900°C increases the hardness of these zones 2-2,5 times. At 1000°C the difference in grain size increases 2 times (7-33 microns) and many of coarse grains have hardness 5,3 Gpa. Structure consists of matrix solid solution with intermetallic needles 200-250 microns in length and 20-30 microns width.

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RELATION OF A COORDINATION NUMBER TO THE SIZES OF FRAGMENTS OF BINARY MIXTURES

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Among programmatic ways, that is at the command of the users of personal computers, large the destiny is made by instruments ways intended for simulation. Computer simulation - method of the solution of a problem to the analysis or synthesis of a composite system on the basis of usage of its computer model.

A series of activities is dedicated to computer simulation of processes of a package, seal and sintering of dispersed stuffs [1-3]. From these activities it is visible, that one of their main characteristics is a coordination number and its value [1].

The actual dispersed mixtures actuate parts of the different forms and sizes, that is they is not monodimension. It in turn influences a coordination number.

There is a number of models, which one allow to determine the characteristics of dispersed stuffs, in particular their coordination numbers. In such models enter parameter of regulation of frame through the characteristics of a surface. Change porosity in model [2] is entered at the expense of parameter of quantity of revolutions of a part to fixation in frame. In other model [3] in adjusting parameter there is a angle α - integral characteristic of properties of parts, which one make embankment. Thus the properties of separate parts are not esteemed.

As against the mentioned models is offered to enter the personal characteristic of a part. This characteristic is the angle of coupling of parts α - angle between a vertical and straight line, that connects centers of parts. Nevertheless, as against [3], at simulation of a concrete stuff the topological characteristics of parts are esteemed, which one influence the factor of the form. The stuff concerns up to one of groups, where α can change in borders of 0-30 grades (part, the form which one close to theoretically round), 30-60 grades (part, form which one deviates from ideal approximately on 50 %), 60-90 grades (part, the form which one considerably deviates from ideal). At simulation of a concrete stuff the angle α in the indicated borders is set incidentally for each part (method Monte-Karlo) [4].

On the basis of the obtained computer model the analysis of relation of a coordination number from a ratio of the sizes of parts of binary mixtures is conducted. In a fig. 1 the outcomes of simulation are submitted.



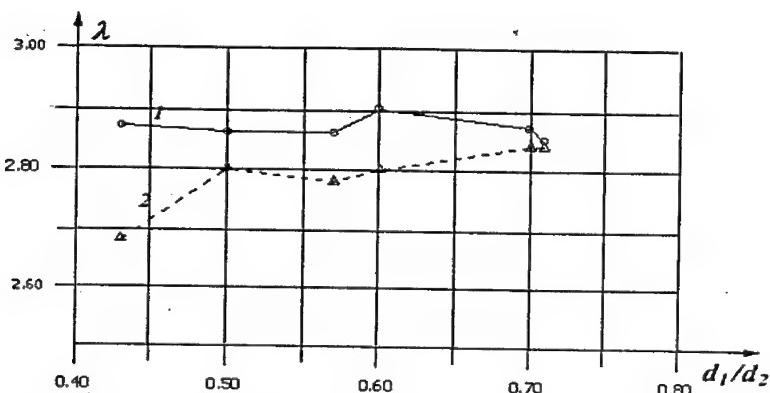


Fig. 1 Relation of a coordination number to a ratio of the sizes of parts of a binary mixture: 1 - kept in repair small-sized fraction in a mixture 50 %; 2 - kept in repair small-sized fraction in a mixture 70 %.

As it is visible from an introduced figure, there is a relation between a coordination number and percentage kept in repair by one of fractions of a binary mixture. At identical quantity of both fractions (curve 1) coordination number is in borders 2,86-2,9. At increase of quantity of a small-sized component in a mixture (curve 2) the coordination number rises from 2,68 up to 2,86 at reduction a difference between the sizes of parts. In both cases at an approaching to monodimension mixture (at increase of a ratio of diameters d_1/d_2) the coordination number follows up to 2,86, that connection with outcomes, introduced in the articles [1].

Thus, the relation of a coordination number to a ratio of the sizes of parts of a binary mixture is obtained.

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DURABILITY OF GALVANIC Cu, Ni AND SCHISTOSE COMPOSITES BASID ON THEM

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The notion of kinetic thermo-Hucuation nature of solid bodies destruction was originally formulated on the basis of studying and analysing the dependence of their durability τ from the Tension σ and Temperature T [1]. At a definite interval σ and T -this dependence is expressed by Dzurkovs equation $\tau = \tau_0 \exp[(U_0 - \gamma \sigma)/kT]$, where τ_0 , U_0 , γ are invariable parameters characterizing solidity of a material property.

The investigation of temperature-power dependence of durability and the analysis of factors of this dependence (especially U_0 , γ) permits to get information about the mechanism of its destruction and intensity of its development.

So, by this time it has been known- that for the metallic materials the magnitude U_0 may take two discrete values, which coincide either with the energy of selfdiffusion or with the energy of sublimation [2] of a metal.

This work is devoted to the results of studying the dependence of τ from σ and T at a one-axle tension in the conditions of creeping for Cu, Ni and for schistose composites Al-Cu and Al-Ni, received by electro-precipitating.

Systematic research of temperature-power dependence of durability of abovementioned materials and an analysis of efficiencies entering this dependence gave the following results:

- As it was awaited, the parameter τ_0 in all cases proved to be the same and within the limits of two orders equal to 10^{-13} c and it coincides with the period of caloritic vibrations of atoms in a cristallic lattice

-The value U_0 (initial energy of stirring up the destruction) for Cu and Ni is accordingly equal to ≈ 170 and 247 KDj/mole which is lower than the energy of their sublimation but closer to the energy value of vacancy selfdiffusion in such materials.

Hence, it is supposed that the mechanism of destruction of the mentioned materials at the analysed intervals σ and T is defined by the diffused development of steams owing to the unbalanced vacancies of a rising origin (when electro-precipitating).



Reasoning in favour of the possibility of diffused mechanism of destruction in such materials is brought in the work [3].

Annealing of the electro-precipitated Cu and Ni at the temperature of close to T fusion has showed that the energy of destruction activation U_0 rises and takes the value which is close to U_0 sublimation of these metals.

-When changing the volumetrical guota of Cu and Ni in the schistose composites Al-Cu and Al-Ni the magnitude U_0 changes unevenly from its value which is typical for Al to the value typcal for Cu or Ni, idest by less values from some volumetrical quota of V* Cu or Ni mechanism of composite destruction is defined by the properties of Al, but by more values from V* properties of Cu and Ni. Our analises has showed that the magnitude V* can be defined under the condition of equality of loading in composite components.

-In all cases of research when changing the state of the material, including the change of volumetrical quota of components, the efficiency γ has been changing and that was a structurully sensitive efficiency which is connected with the exponent of local tension level.

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COMPUTER SIMULATION OF THERMOEXERTIONS IN HEATPROOF LAYERED COATINGS

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Creation of heatproof multilayered coatings for the protection of surface of details from hightemperature corrosion takes on special significance at development of new generation of energy-power plants (gas turbine, etc.) Thermal exertions between a base and coating and also between separate layers in coating define adhesion and mechanical strength of the coatings.

In order to provide a optimal exertion state in the layered system, we have built the mathematical model of the multilayered coating based on calculation of mechanical exertions in the layers arising from heating. The model provides calculation of the multilayered coating including different materials, coating deposition temperatures and geometric characteristics. The model based on mechanical equilibrium equations of the system taking into account thermoexertions

$$\begin{cases} \frac{\sigma_i}{E_i} + \alpha_i \Delta T = \frac{\sigma_{i+1}}{E_{i+1}} + \alpha_{i+1} \Delta T, i = 1, n-1 \\ \sum_{i=1}^n \sigma_i d_i = 0 \end{cases} \quad (1)$$

where

σ_i — exertion in i -th layer	d_i — thickness of i -th layer
E_i — Young's modulus of i -th material	ΔT — temperature difference
α_i — thermal expansion coefficient of i -th layer	

Test calculations were made for 2- and 4-layered coatings of yttrium chromite (YCrO_3) dopped by (8% Y_2O_3) zirconium dioxide for the system of $\text{Cr}-(\text{YCrO}_3-\text{ZrO}_2)\cdot n$, $n=1,2$. Thickness of the layers is equal and ranges within 0,5 - 10 mkm.

The calculations were shown that doubling of number of layers in the $\text{YCrO}_3-\text{ZrO}_2$ system does not practically influence on exertion in every layer. The exertions in the 4-layered coating remain practically the same as for the 2-layered coating. Thermoextentions depend on temperature and the ratio of thickness of layers in the coating.

The results of calculation shows that tensile exertions in YCrO_3 layer at temperature of 1473 K increase from 99 MPa to 151 MPa at changing the ratio of thickness of $\text{ZrO}_2/\text{YCrO}_3$ layers from 1 to 10. It essentially exceeds the limit of strength of YCrO_3 (50-70 MPa). At the thickness ratio of $\text{ZrO}_2/\text{YCrO}_3$ layers lower



than 1, thermoexertions in YCrO_3 layers decrease and at the ratio of 0.25 and 0.1 are 46 and 22.4 MPa accordingly that do not exceed the limit of strength of YCrO_3 .

Thus, the calculations of thermoexertions in the 4-layered coating like $\text{YCrO}_3-\text{ZrO}_2$ have shown that this pair has a low level of thermoextentions at some conditions and can be used as a protective barrier for Cr and its alloys from diffusive gas saturation at high temperatures.



FEATURES of STRUCTURAL CHANGES in a CHROMIUM-COVERING LAYERED COMPOSITIONS during HIGH-TEMPERATURE PROCESSINGS

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High-temperature using of super chromium alloys is interfered by their quite sensitive to gaseous impurity and surface defects, which do their embrittlement. The application of coverings is one of basic (after alloying) ways of increasing of capacity for work of chromium alloys. For maintenance of protection from oxidation and high-temperature corrosion of a covering should have phase and thermal stability.

In work the influence of parametres of high-temperature processing in various environments (argon, oxygen, nitrogen, air) on porous, phase structure and microstructure of a protective heat resisting covering on a basis of ittrium chromite is investigated.

Four types (YCr, YCrO, CrO - YO, YCr + YCrO) coverings received by various technological parametres of condensation with ion bombardment (CIB) are investigated.

Layered oxide covering (CrO-YO) and the coverings on a basis of ittrium chromite (YCr + YCrO) are transformed already during forming annealing (Ar, 1473 - 1673 K), and, if the layered coverings undergo significant changes, the coverings of system Y - Cr - O in the greater degree keep the composition. The basis direction of transformation of type YCrO and YCr + YCrO coverings - growth porosity and change drop phase. The rise of temperature of forming annealing in argon from 1473 up to 1673 K during 2 h results in increasing open porosity from 4,2 % up to 6 % and reduction drop phase from 6,1 % up to 2,5%.

It is shown, that under annealing of the formed coverings in air, structure become rough for all types of coverings.

The microstructural researches of coverings have allowed to reveal connection features of structural changes under annealing and mechanical properties of a substrate material. Chemical-thermal treatment (CTT) in oxygen atmosphere ($T = 1323$ K during 6 h) results in reception petty grain structure, the size of a grain in a covering decreases from 0,6 up to 0,065 microns. Under annealing in air (1573 K, during 6 h) are revealed a plenty round pores of the sizes from 0,007 up to 0,2 microns and separate formation of the size ~ 0,01 microns located, basically, on grain boundaries.

However coverings on samples are effective barriers on the nitrogen diffusion ways in chromium, therefore the structure of layers near surface of a substrate changes. After high-temperature annealing in air (1473 K, during 115 h) in



structure are absent continuous nitride layer and nitride net and the microhardness in layers by thickness up to 40 - 50 microns grows in 1,5 times, at the same time for samples without a covering the microhardness is increased in 5 times, and the thickness nitride of a layer reaches ~ 100 microns, which is responsible for embrittlement of a substrate material.

Thus, with the help of the carried out researches is established, that the structural transformations in a layer of a covering at high-temperature annealings are responsible for a level of barrier properties of coverings and mechanical properties of a substrate.



**PROCESSING PARAMETER INFLUENCE ON STRENGTH AND
STRAIN PROPERTIES OF TEG-BASED CM INFILTRATED
WITH PYROLITIC CARBON**

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Thermally expanded graphite (TEG) products have potential due to high temperature, radiation and corrosion resistance, recovering and compressibility and room temperature formability at low pressures without binder use [1, 2]. Gas-phase infiltration of a porous frame made of TEG with pyrolytic carbon is one of methods of production of composite materials (CM) based on TEG that have elevated service properties.

In this work to prepare TEG particles, a technology was used that involved treatment of graphite powder with sulfuric acid and oxidizer, and thermal expansion of oxidized graphite. Parts in the form of rings were used as porous medium made of TEG by unilateral pressing [2]. Strength and strain properties of TEG-based CM with a density of frame varying between 0.07 and 0.5 g/cm³ and different frame porosity were studied.

To perform infiltration with pyrolytic carbon, we used a technology developed by National Scientific Center, Kharkiv Physical & Technology Institute, that employed a radially moving pyrolyzing zone [3]. A rod with impaled rings as frames made of TEG were resistance heated in a flow of natural gas (CH₄) up to a temperature T_K. The pyrolytic reaction of TEG deposition occurred only within a narrow zone of pyrolysis around the rod. The required degree of densification for composite material was achieved due to a high affinity of graphite and pyrolytic carbide deposited. For example, initial density of rings non-infiltrated with pyrocarbon was 0.07 g/cm³, whereas infiltrated ones showed an apparent density between 0.3 and 0.5 g/cm³.

Microstructures and macrostructures of TEG frames were investigated as fabricated and of those infiltrated with pyrolytic carbon. Results of fracture examination revealed a uniformity of saturation with pyrolytic carbon for TEG frames of different densities.

Strength/strain properties of specimen rings were investigated at axial compression and stress-strain diagrams were registered. The TEG frames density and infiltration completeness with pyrocarbon were also examined for the effect upon the CM mechanical properties. It was found that the gas-phase technology considerably increased the modulus of elasticity at compression E_C, strength at 10% deformation, σ₁₀, and varied elongation, ε_B, in comparison with specimens of similar densities but non-infiltrated with pyrolytic carbom. Thus, modulus of elasticity at compression of rings with the density of 0.5 g/cm³ E_C increased 8 to 10 times (from 10-15 to 80-120 MPa), strength at 10% deformation, σ₁₀, 3.5 to 4.5



times (from 1.0-1.2 to 4.0 to 4.5 MPa), elongation, ϵ_B varied from 8 -15% to 40-45%. Similar data were obtained for the rings with apparent density of 0,3 g/cm³.

As the study of composite materials of TEG-pyrolytic carbon system has shown, the said materials combined a low density and rather high loading ability, and can be used in various industries.

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A NEW LAYERED STRUCTURAL COMPONENT IN Fe-C ALLOYS

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In Ref. /1,2/ description was given to a structure in Fe-C alloys treated by a laser beam pulse with a high microhardness about 9 Gpa. Recently it has been discovered also on the surface carbon steel submitted to electrospark erosion. It was called "baikovite" in memoriam of academician A.A. Baikov, who in 1910 proposed to consider cementite as carbide phase which becomes non-stoichiometric (Fe_3C_{1-x}) at high temperatures with a deficit "x" in carbon. Such a phase is able to decompose upon cooling into an ultrafine mixture of stoichiometric Fe_3C and an excess of austenite in the form of very thin strata mainly along the (002) planes of the iron-enriched cementite. In this respect "baikovite" is analogous to secondary cementite platelets in hypereutectoid steels /3/.

Yet such a simple concept has some uncertain aspects:

1. Carbon concentration in Fe_3C_{1-x} must be much less than 6,4 mass % /4,5/ because of the rather large amount of austenite strata in "baikovite" (see fig.) These strata have many excrescences seemingly of secondary nature, deposited as extra γFe during cooling of metal.

2. "Baikovite" and ordinary ledeburite co-exist only in a very narrow band between these two types of structure (see fig.). Away from this thin strip ledeburite cementite does not contain no strata of "baikovite". From this we can deduce that the honeycomb eutectic austenite pillars (martensite + retained austenite "islands" in fig.) must have been formed from molten metal. Then maybe molten partly decarburised cementite remained during a very short lapse of time (milliseconds) in the form of "liquid crystals"?

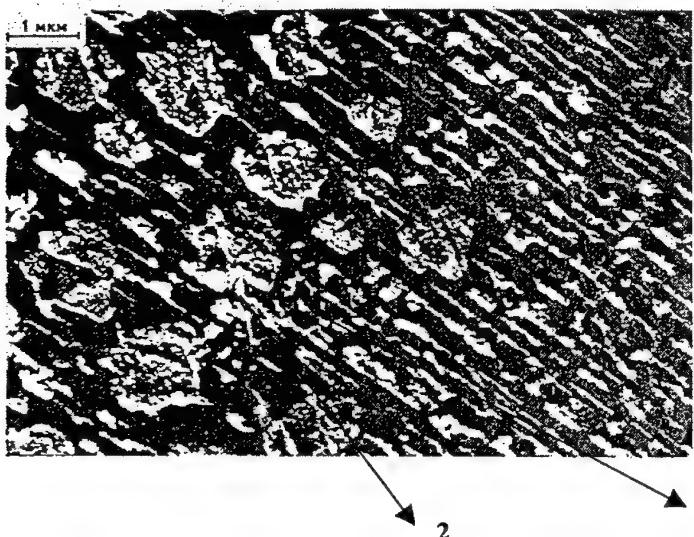
This can explain the formation from such a cementite of the textured structure of baikovite with strictly parallel thin layers of excess austenite. Yet such a hypothesis must be refuted upon the following grounds. In figure we can see that in the vicinity of eutectic austenite pillars the amount of excess austenite strata quickly diminishes. At places there is no such layers at all in the nearest environment of austenite "islands". Probably here it is more "easy" to excess iron to be deposited on priorly formed γFe . And this means that baikovite is formed *a posteriori* in a solid phase reaction. Therefore the "liquid crystals" route seemingly should be discarded.

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Caption under figure

TEM microstructure (after chromium shading) of baikovite strata on the borderline between baikovite and ordinary ledeburite, whose honeycombs are seen as "islands" with a structure of martensite and retained austenite (at room temperature). Nital etched. X24000

1. The thinnest plates of baikovite austenite;
2. Cross-section of "columns" of ledeburite austenite.



THE MULTILAYER METAL-CERAMIC COMPOSITES, FABRICATION AND PROPERTIES

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Wide using of advanced ceramic materials, which have high hardness, wear resistance, mechanical strength, is restricted by it high brittleness. The shock resistance of ceramic can be raised and the development of cracks can be stopped in particular by introducing of metal part, which have high viscosity in material.

Metal-ceramic composite can be obtained by different way, in particular: by introducing of metal component in ceramic matrix as particles or fibre (wire); by layer to layer joining of ceramic and metal plates. Ceramic and metal plates can be joined by gluing or brazing. The brazing method has same advantages.

The multilayer macrocomposite and the way of it obtaining were chose in this investigation.

As the ceramic component B_4C+ZrB_2 , Si_3N_4 , $TiN-Al_2O_3$, $ZrO_2-Al_2O_3$ systems were chose.

As the metal component different materials (copper, titanium, kovar) were used.

The wetting of ceramic by metal melts was studied, the optimal composition of filler for brazing of ceramic materials was selected.

The technology of obtaining of multilayer metal-ceramic composites with layers thickness of 1, 1.5, 2 mm for ceramic and 0.5, 1, 1.2 for metal in different combinations was elaborated, the quantity of layers ranged from 5 to 10, maximal dimensions of plates were 150×150 mm

The mechanical and thermal properties of such composites were studied. The bend strength of composite in perpendicular to layers plane is equal and in parallel to layers plane is equal or 20-50% higher to that of matrix ceramic. Shock resistance of composite 1.5 – 10 times higher than that for matrix ceramic of the composite. The anomaly of the composite thermal expansion was found — in perpendicular to plates direction the TEC is signifintly higher then that calculated using additivity rules.



THE STUDY OF POWDER LAYERED MATERIALS BEHEVIOR DURING THE PURE BEND TESTING

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The wear-resistant, corrosion-resistant and tool materials are the most interesting among the powder layered materials. Such materials are produced by joining of layers of high wear resistant alloys with ones on the more cheap and more deficit base – carbon steels, alloyed steels or carbide-steels.

It allows at the same time to improve the physical, mechanical and tribologic properties of articles and substantially to decrease their cost price due to economy of the deficit components which are used for producing of the work layer.

The materials of the such types as of composites on the base of the stainless steels, the hard metals-steel, layered hard metals and sintered the soft-magnetic materials having layered structure are the most bright samples of the layered materials to day.

The physical and mechanical properties of layered materials were investigated with mathematical modeling and method of pure bending testing.

The analytical expressions for determinating the effective elastic modulus and maximum strengthes at wich the destruction of the top layer is beginning during the change of the work layer thickness in the limit 0-50% of the all hight of the layered specimen.

This expressions are written by such means that only one variable is clarified in accordance to the experimental date.

The difference between the property values, which are obtained with the mathematical modelling and experimental values, measured on the specimens in the form of two-layered or symmetrical three-layered simple beam, don't exceed 10% that corresponds to the measurement error.



Distruction of coatings on titanium base by cycle loading¹⁵⁵

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The aim of this work is investigation of influence refractory coatings on resistance of contact fatigue, wear and fatigue crack broadening on different methods of coating plotting.

Electrospark treatment work out on EFI-46A plant, vacuum electrospark plotting on 01NN plant. Diffusion deposition work out by contact method in powder mixture in containers. After this treatment work out electroultrasonic treatment on turning plant with use of ultrasound generator UZDN-1.

Diffusion titanation of steel pass with formation of zones on specimen depth. Each of zones has determined concentration of titanium and concrete phase composition. Thickness of hardened layer increase, hardness of surface layer some decrease and increase the wear resistance. Preliminary hydrogenization of steel increase antifrictional properties of steel and hydrogen intensify diffusion processes. The steel after diffusion titanation has more resistance to contact fatigue in contrast to steel 3 with electrospark coating. This connect with increasing of hardness of specimens core after diffusion titanation. Cycle contact loading of this specimens (with coating $H_{\mu}=15-16$ GPa and soft base) leads to exfoliation of coating.

Complex treatment (titanation + hydrogenisation + electroultrasonic treatment) leads to tension residual stresses formation in surface layers. Residual stresses spread to depth in 4-5 times more than coating depth. Such treatment increase resistance to cycle crack steadfastness.



¹⁵⁶ Spark Treatment of Steel by TiN-Mo-Ni Materials

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Molybdenum have high erosion resistance, sufficient heat resistance and nonsufficient steadfastness to oxidation. Molybdenum form hard solution with iron and molybdenum is useful addition to refractory titanium nitride base by creation of new electrode material.

Phase diagram TiN-Mo is absent in literature although it can be proposed that this system is quasi binary eutectic one. In connection with this it is interest to study the influence of molybdenum content in titanium nitride composition on spark treatment process characteristics. Spark treatment work out in EFI-46A plant in regime I=1,5A, U=15V. In this work studied dependence of sum signification of anode erosion and cathode overweight on dependence of treatment time.

Anode erosion and cathode overweight decrease with increasing of molybdenum content in contrast with pure titanium nitride erosion. 20% content of molybdenum addition to titanium nitride is optimal.

High molybdenum content leads to increasing of friction coefficient, heightened warming up and grip of working surfaces of hardened details.

However addition of molybdenum not leads to sufficient changes in surface formation. Investigation of Mo-Ni system show that optimal formation of layer occur by 40%Ni + 60%Mo composition.

MoNi intermetallide formation leads to appearance cycled structure in electrode material. This structure is careful to layer formation and wear resistance of surface.



STRUCTURE AND PROPERTIES OF CARBIDES COVERINGS ON STEELS AND FIRM ALLOYS

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In work on the basis of the analysis of physic-chemical conditions saturation there were reception results, which have allowed correctly to approach to development of new ways of drawing carbide-forming by elements, to define parameters of process of drawing of coverings with forecast by structure and properties.

Carried out classification of sating materials by character of influence of initial structure on thickness, microhardness of coverings, classification of coverings by phase structure and character of an arrangement of layers.

The influence of parameters of process and sated material on structure, textured of the characteristic, chemical structure of coverings is established.

Carbides of a covering were rendered in the closed reactionary space at the lowered pressure of a gas phase with use as initial components of powders of transitive metals and four-chloride carbon.

The saturation of tungsten BK8, TT8K6 and tungstenfree firm alloys TH-20, KHT-30 and KXH-15 with addition in reactionary space at last stage of process, simultaneously, nitrogen and oxygen was carried out. In result on a carbide surface of a covering it turned out oxide and nitride blanket, which together with titanium carbide formed complex oxide – carbide and nitrogen – carbide of connection.

As initial components at reception carbides of coverings on steel tool the powders of titanium, chromium, vanadium, boron carbide, and mix of niobium – chromium were used.

With the purpose of definition of tribometries properties of coverings the tests, «simulating» that is close to real conditions, by the machine of friction MT-68 were carried out(spent), which design permits simultaneously: independently to measure force of friction (factor friction), deterioration of a sample and temperature in a zone of contact.

As a result of realisation of experience are revealed decrease factor frictions at «dry friction » twice, and also reductions of wear process at 8-25 time at comparison of samples with covering and without.



RESISTANCE OF LAMINATED CARBON FIBER-REINFORCED PLASTICS TO DEFORMATION AND FRACTURE AT ROOM AND LOW TEMPERATURES

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Regularities of deformation of laminated carbon fiber-reinforced plastics and their strength have been studied at room and low (above 77 K) temperatures. On the basis of the energy smoothing method of V.V. Bolotin the laminated carbon fiber-reinforced plastic with a symmetrical scheme of laying fibers with respect to its median surface is regarded as an orthotropic body. From the tensile, compressive, shear and bending tests all characteristics of elasticity and strength of an equivalent orthotropic body have been determined. By testing specimens with one direction of reinforcement similar parameters have been obtained for an individual monolayer. The influence of different experimental procedures on the accuracy of the evaluation of the rigidity and strength characteristics of the laminated carbon fiber-reinforced plastic has been investigated.

The efficiency of the calculation procedure for the elastic deformation of a monolayer has been analysed on the basis of the mechanical characteristics of its components: carbon fibers and an epoxy matrix. The authors suggest an approach for the evaluation of the characteristics of elasticity and fracture stresses of a laminated composite material.

The studied carbon fiber-reinforced plastic with the symmetric scheme of laying layers with respect to its median surface is shown to have a considerable difference in the values of the elasticity characteristics in the direction of the principal axes. The parameters of elasticity of the material remained practically unchanged when lowering the test temperature down to 77 K. The carbon fiber-reinforced plastic considered features low resistance to interlaminar shear. The maximum shear stresses are 50-60 times lower than the maximum normal stresses depending on the scheme of the layer arrangement.

Lowering the operating temperature of the carbon fiber-reinforced plastic down to 77 K results in the reduction of its shear and tearing strengths approximately by 3-30 %. The single cooling of such composite to the temperature of liquid nitrogen has a hardening effect (up to 10-12%) on its strength if then it is operated under room conditions. In further temperature cycling the strength characteristics of the carbon fiber-reinforced plastic remain unchanged.



SHAPE MEMORY MATERIALS

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PREDICTION OF STRAIN HARDENING OF METALLIC MATERIALS
UNDER COMBINED CYCLIC LOADING

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Experimental investigations into low-cycle fatigue of metallic materials under strain-controlled loading demonstrated that cyclic instability of mechanical characteristics within the first 10-30 cycles is common to the majority of materials. As a rule, under such conditions cyclic instability manifests itself in the strain hardening progressing from cycle to cycle with a decreasing speed until it reaches the state of complete stabilization of cyclic properties of the material.

The level of strain hardening is determined by the maximum value of equivalent stresses. Depending on the complexity of a cyclic path at the same level of the maximum equivalent strain metallic alloys are characterized by different levels of strain hardening. The lowest level of hardening corresponds to proportional cyclic paths while the higher level corresponds to nonproportional ones. Low-cycle fatigue testing showed that the level of strain hardening essentially affects the service life of the material. The higher the stress level in a stabilized state the shorter is the service life of the material under cyclic load.

The paper considers the effectiveness of the prediction of strain hardening for chrome-nickel steels during low-cycle fatigue under conditions of the nonproportional biaxial deformation by tension - compression and alternating torsion.

Four groups of the experiments (taken from the literature) have been analyzed with a total number of cyclic paths being 20. The coefficient of cycle nonproportionality, which characterizes the geometry of the deformation cycle, has been employed as a governing parameter. Comparative analysis has been performed of the effectiveness of the use in calculation practice of the four most known in the literature coefficients of nonproportionality of a cycle for the prediction of the maximum level of strain hardening.

The most effective coefficient has been shown to be the parameter that is plotted with allowance made for the following three factors: the effective area enclosed by the cyclic path, the cyclic path length, and the path orientation with respect to the direction of the action of the maximum principal strain.



THE SHAPE MEMORY EFFECT AND SUPERELASTICITY IN SINGLE CRYSTALS OF TITANIUM-NICKEL

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On single crystals Ti-30%Ni-20%Cu(I), Ti-40%Ni-10%Cu(II), Ti-50,2%Ni(III), Ti-50,6%Ni(IV), Ti-51%Ni(V), Ti-51,5%Ni(VI) (at. %) systematical investigation of mechanical properties of B2 phase, the shape memory effects and superelasticity in dependence on type of martensite transformations, crystal orientation, sign of applied stresses (tension/compression), size, volume fraction and number of crystallographic variants of dispersed particles Ti_3Ni_4 has been carried out.

1. B2-B19 martensite transformation is observed in crystals (I). Values of superelasticity and shape memory effect, its dependence on orientation and sign of applied stress are agreed with theoretical values obtained at calculation of deformation of lattice only. Therefore, B19 martensite single crystal arises at deformation at $T < M_s$ and passes to single crystal of B2 phase at heating at $T > A_f$.
2. B2-B19-B19' martensite transformations are observed in crystals (II) moreover B19-B19' transformation turns out to be not complete. Consequently, final product of transformation is the mixture of two martensite phases - B19 and B19' - at cooling at $T < M_s$ and at deformation at $M_d > T > M_s$. For the first time experimental data on orientation dependence of shape memory effects and superelasticity were obtained. It has been found that volume fracture of B19' phase appearing under loading is dependent on crystal orientation and deformation modes.
3. In quenched crystals (III-VI) B2-R-B19' martensite transformations are observed. Values of shape memory effect are dependent on orientation and are agreed with theoretical ones obtained at calculation of lattice deformation only. Superelasticity is only observed at compression in [001] in which slip deformation of B2 phase $a < 100 \{110\}$ does not take place because Schmid's factors is equal to zero. The main deformation mechanism in B2 phase is mechanical twinning.
4. Aging of crystals (IV-VI) without loading leads to formation of four crystallographically equivalent variants of particles Ti_3Ni_4 . It changes mechanical behavior of these crystals on compared with single phase ones on principle. It has been experimentally found that R-B19' martensite transformation progresses by two steps in crystals with particles Ti_3Ni_4 of 400 nm in size. At first transformation begins near the particles and only in further



cooling it occurs in the places of crystals located far from particles. In aged crystals superelasticity appears in all orientation at compression and tension which is absent in quenched specimens. Values of shape memory effect and superelasticity are determined by size and volume fraction of dispersed particles Ti_3Ni_4 and decreased with increasing of volume fraction of particles. It has been shown that the temperature interval of superelasticity can change from 30K to 150K in dependence on size particles. In single crystals of Ti-Ni containing dispersed particles the loops of superelasticity have asymmetrical form at loading - unloading and the decreasing of mechanical hysteresis with increasing of test temperature are observed.

5. Aging under loading of <111> and <110> single crystals leads to the formation respectively of one and two variants of dispersed particles Ti_3Ni_4 . The values of shape memory effect and superelasticity in crystals containing one or two variants of dispersed particles are lesser than in crystals with four variants. The shift of points of martensite transformation to range of high temperature, the decreasing of critical resolved shear stresses at $T=M_s$ and the decreasing of temperature interval of superelasticity are observed in aged crystals under loading.
6. The micromechanical model of the shape memory effects and superelasticity in crystals containing dispersed particles was elaborated. It has been shown that main moment in control of functional properties - shape memory effects and superelasticity - in aged single crystals of titanium-nickel is mechanism of interaction of martensite crystals with dispersed particles and the meeting condition of compatibility on interface «martensite - particle». The model describes the dependence of shape memory effect and superelasticity on parameters of microstructure of single crystals TiNi at aging.

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EFFICIENCY OF A SPRING DEVICE FROM TiNi SHAPE MEMORY ALLOY

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The unique physical-mechanical properties of materials with a shape memory effect allow using them in different devices for work performance. In particular, the springs produced from TiNi alloy ideally applicable for thermo-controlled actuators which capable to give significant displacement and to generate considerable stress.

In the present work an efficiency of the spring device produced from TiNi shape memory alloy was studied in conditions of heating by electro-current on air. Such mode represents near-actual conditions of operation of real actuators. Note that in the majority of the known publications the origin of stress and displacement of actuators with a shape memory are explored at different stiffness of counteraction. In this paper examination of the actuator behaviour is undertaken at the variation of stiffness of a working body with a shape memory effect.

The experimental set up represented two sequentially joint springs: steel and fabricated from a wire of TiNi alloy by 1,91 mm diameter. A diameter of coils of a TiNi wire was varied. Thus stiffness of a spring was changed. The stiffness of a steel spring was equal 33,6 H/m. The opposite ends of linking were fixed. Measuring of strain realized by the pointer fixed in a place of linking of springs. A spring from TiNi alloy heated by AC passed through the secondary circuit of the transformer. A current value was changed by variation of voltage in a primary circuit. In the martensitic condition, the shape memory spring was stretched, and steel spring was unstressed. At heating the shape memory spring compressed, thus a steel spring was stretched. After the current was turned off the shape memory spring stretched up to tentative length at the expense of elastic force of the steel spring. The work done by the device was calculated on value of elastic energy of a steel spring.

The experience have shown that the displacement in the device with less hard spring by a diameter 3,5 cm lags the displacement in the device with a hard spring by a diameter 2,5 cm (fig. 1). Notes the temperature of the shape memory working body in selected experimental condition is proportional to power P. Therefore the temperature kinetics of stress generation at martensitic transformations in TiNi depends not only on stiffness of shape recovery counteraction but also from stiffness of the most working body as it follows from obtained data. Obviously, it is necessary to take into account a relation of stiffness of interacting springs when



actuator is developed

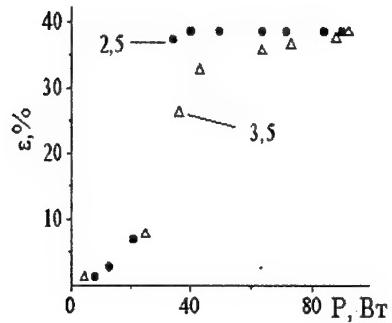


Fig. 1 Dependencies of the spring strain on electrical power in a secondary electrical circuit. The markers specify the diameter (in cm) of TiNi spring coils

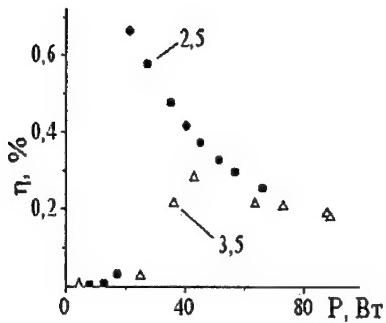


Fig. 2 Dependencies of the relationship between mechanical work done and electrical power in a secondary electrical circuit, from the value of this power. The labels are similar to a fig. 1.

Efficiency of system consisting of fissile and passive devices has a maximum (fig. 2). The best efficiency of electrical energy transformation is reached for a spring with greater stiffness as it can be seen from figure. The maximum of the efficiency is on the initial stages of strain recovery when the rate of strain change is great. Hence, the shape memory actuators should be constructed so that the necessary displacement was realized on the partial temperature interval of martensitic transitions.

The results obtained in the present work concern with the initiation of the one-way shape memory effect. Obviously, at cycling working of the actuator, other peculiarities and essential dependencies of energy transformation efficiency on number of thermal cycles will be observed.



HEAT FLUX IN TiNi AT THERMOCYCLES UNDER CONSTANT LOAD

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Heat flux and torsional deformation at cooling and heating of $Ti_{55}Ni_{44.5}Fe_{0.5}$ alloy specimens loaded by a constant torque have been studied. The applied tangential stress was varied between 0 and 80 MPa. The tube-shaped specimen had the dimensions: inner diameter $d_{in}=5.9$ mm, outer diameter $d_{out}=4.9$ mm, length $L=24$ mm. Preliminarily the specimen was quenched from 800 °C and then aged at 550 °C for 3 hours.

Fig. 1 demonstrates the result of the differential thermal analysis (DTA) — the dependence of the difference ΔT between the specimen and the copper reference sample temperatures on the specimen temperature. At cooling and heating one can see anomalies on the curve, which correspond to the martensitic transformations $B19' \rightarrow B2$ and $B2 \rightarrow B19'$.

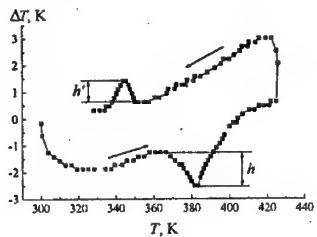


Fig. 1. DTA curves at heating and cooling without stress.

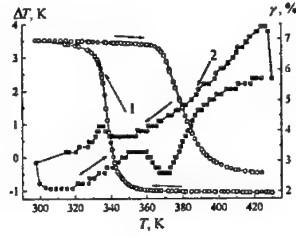


Fig. 2. Dependence of strain on temperature (curve 1) and DTA (curve 2) at cooling and heating under shear stress 60 MPa.

Fig. 2 presents the results of one of the experiments on simultaneous measurements of the deformation (curve 1) and heat flux (curve 2) at thermocycling under the stress $\tau=60$ MPa. The anomalies on the curves 1 and 2 do not coincide. The strain recovery at cooling starts before the beginning of $B2 \rightarrow B19'$ transition. Previously the same results were observed in [1]. The strain recovery at heating starts when the martensitic transformation is already in progress (the minimum of ΔT on curve 2). By contrast, in [1] the strain recovery at heating was observed before the start of martensitic transformation. The authors connected this phenomenon with twinning. In our case twinning was not observed,



evidently because the specimen was well annealed (it did not demonstrate the $B2 \rightarrow R$ transformation at cooling).

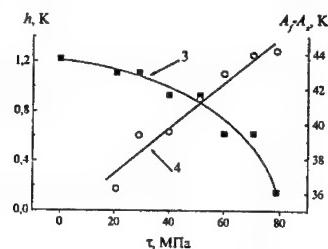


Fig. 3. Dependence of DTA peak height (curve 1) and the temperature range of transformation ($A_f - A_s$) on the applied stress.

It was found that when the stress is increased the values h and h' of anomalies of curve 2 (fig. 1) decrease. The dependence of $h(\tau)$ is presented in fig. 3 (curve 1). Curve 2 in the same figure illustrates the widening of the temperature interval of the reverse transformation ($A_f - A_s$) when the stress is increased. Probably, this widening is the cause of the observed decrease of the DTA peak height. Besides, the latent heat of the transformation may decrease under the stress. This follows from the results described in [3], in which it was observed that under the stress the $B19' \rightarrow B2$ transformation may acquire the features of the 2nd order phase transformation.

From the data obtained in this work one may conclude that:

1. in well-annealed TiNi-based alloy the strain recovery is due exclusively to the transformation (not to twinning);
2. stress may be responsible for the decrease of the latent heat of the transformation.

Acknowledgments

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DURABILITY OF SHAPE MEMORY EFFECT

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The experiment shows that the shape memory effect (SME) while in service in nickel and titanium alloys fades out before destruction. Therefore, resource of work of these alloys can be determined by reduction and disappearance of their functional properties [1]. It can be connected to generation of defects and their evolution during repeated of martensite transformations.

The main goal is to research of evolution of an elastic field of defects during repeated of martensitic transformation in the equiatomic TiNi alloy. The research was carried out at a structural microlevel using x-ray method, which allows to distinguish martensite from austenite.

Experimentally, martensitic transformation was realized by cooling of the beam from nitinol and bending, then was warmed up to of a austenite condition. In process of heating the beam was straightened accepting the primary form. Such thermo-mechanical loading was executed cyclically.

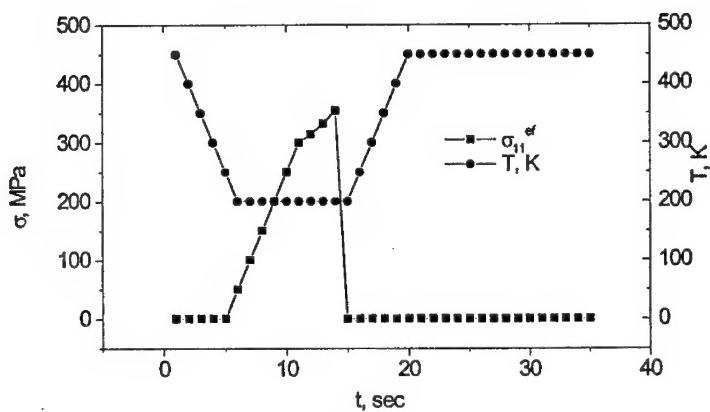


Fig. 1. The scheme of thermo-mechanical loading.



In the figure 2, changes of the shape of a x-ray line in process of thermo-mechanical loading are presented.

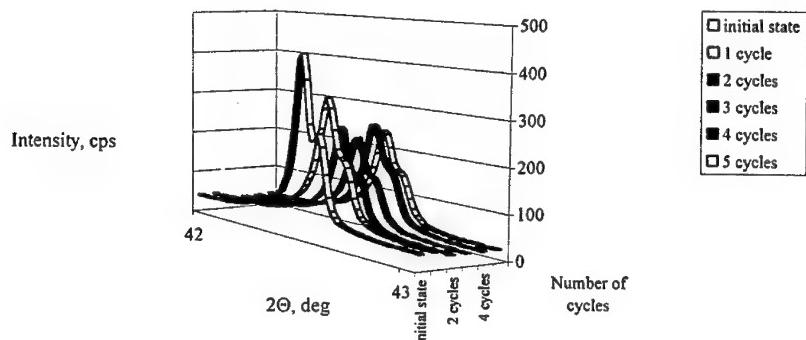


Fig. 2. Changes of the shapes in the austenite x-ray line 110 during loading.

Applying x-ray method and harmonic analysis have been determined the changes of structural parameters (Fig. 2).

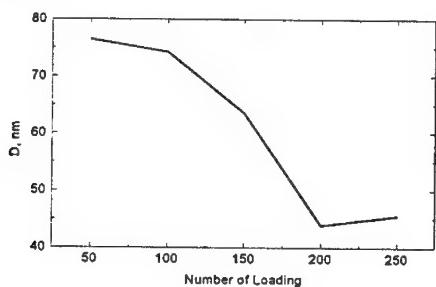


Fig. 2. Changes of size of koherent blocks during loading.

The received experimental results are presented of the base of computer modeling. On the basis of these results it is represented, that the parameters of thin structure can be used as determining parameters of durability SME.

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KINETIC OF RADIATION DAMAGE OF TiNi ALLOY BY REACTOR NEUTRONS

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In previous works of the authors it has been shown that the temperatures of martensitic transitions in TiNi shape memory alloy decrease exponentially when the neutron fluence increases, this process starting and going on intensively since the very low fluence when irradiation start [1]. The nature of the radiation damages in TiNi, causing such sharp changes of the kinetics of phase transitions, is still unclear. There is an idea that the discovered regularities are connected with disordering of the crystalline lattice of the ordered alloy.

The present investigation has the goal to study the kinetics of neutron irradiation damages in TiNi alloy.

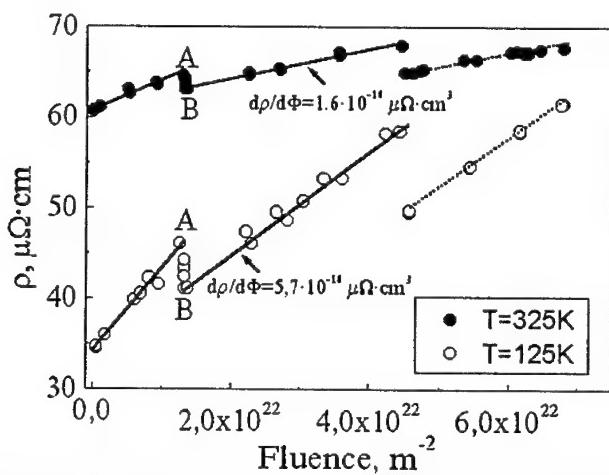
Irradiation was carried out in the low-temperature helium loop of the WWR-M reactor at PNPI RAS. Irradiation up to the fluence of $5.5 \cdot 10^{22} \text{ m}^{-2}$ was at 120 K (in the martensitic state of the alloy). After this the temperature was raised up and the further irradiation was conducted at 330 K (austenitic state). The damages were judged by the variations of electric resistivity. To take the dependencies of resistivity on temperature the specimens have been given thermocycles in the interval 120 – 350 K without stopping the irradiation

Fig.1 shows how the resistivity at 125 K and 325 K depends on the fluence of fast neutrons ($E > 1 \text{ MeV}$). Solid lines present the results obtained at low-temperature (120 K) and the dashed ones – at high-temperature (330 K) irradiation. At point A the reactor was stopped, with the temperature growing up to $\approx 340 \text{ K}$, and at point B the irradiation was restarted at $T = 120 \text{ K}$. One can see on the figure that the rate of the resistivity growth $d\rho/d\Phi$ in the austenitic state of TiNi alloy is very high and exceeds by one – two orders the values of $d\rho/d\Phi$ measured in pure metals at low-temperature irradiation [2]. But, the resistivity of martensite phase grows 3 times faster.

One can explain the high value of $d\rho/d\Phi$ in the austenitic phase by the decrease of the long-range order due to irradiation. The total growth of the resistivity in the austenitic state can be presented as the sum of the two contributions: $\Delta\rho_v$ caused by the growth of vacancies concentration and $\Delta\rho_d$, caused by disordering: $\Delta\rho_{aust} = \Delta\rho_v + \Delta\rho_d$. An additional scatter of charge bearing particles in the martensitic state may be connected with the increase of the density



of interfaces between the domains of martensite belonging to different crystallographic orientation. The refinement of the martensite polydomain structure provides an additional contribution to resistivity $\Delta\rho_b$. By this means $\Delta\rho_{\text{mart}} = \Delta\rho_v + \Delta\rho_d + \Delta\rho_b$.



Resistivity versus fluence for TiNi alloy irradiated at 120 K (solid lines) and 330 K (dot lines).

Thus, in the present work new data concerning the kinetics of accumulation of irradiation damages in TiNi alloy with martensitic transformations has been obtained. It has been shown that the electric resistivity in the martensitic and austenitic states increases with different rates. Contributions of several structure processes into the total variation of electrical properties have been considered. It has been established that annealing of irradiation defects leads to the complete recovery of the resistivity and to the restoration of the temperature kinetics of martensitic transformations.

This work was supported by the program "Neutron investigation of condensed matter" (contract N 107) and by the program of RFBR for leading scientific group support (grants N 00-15-9627, 00-15-96023)

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THE INFLUENCE OF AGING UPON REVERSIBLE SHAPE MEMORY EFFECT IN TITANIUM NICKELIDE

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Usually one can connect reversible shape memory effect (RSME) with the effect of internal strain fields created by dislocations. The aging takes place in TiNi alloys with the nickel concentration more than equatomic composition. The Ti_3Ni_4 phase precipitation essentially influences on the following martensitic transformations. The object of our investigation was the study of an influence of aging upon the austenitic and martensitic type RSME in the titanium nickelide. The Ti-50,5 at.%Ni alloy plates were made by rolling in the austenitic state (in an industrial conditions). The investigation of austenitic type RSME was carried out after rolling and after tension of quenching samples until 6% at 350°C. Martensitic type RSME was investigated after tension of quenching samples until 12% at room temperature. The temperatures of martensitic transformations and the deformation in RSME were defined by dilatometric method.

The aging Ti_3Ni_4 particles with approximate length 70 nm was found after rolling in austenitic state. The particles was arranged orientatively, their long axis has one preferable direction. Thus the aging process goes during rolling and cooling and with the texture of a deformation of B2-phase the texture appears in particles arrangement. Dissolving of the particles Ti_3Ni_4 and poligonization B2-phase takes place after heat treatment at 500°C. The aging goes again in time of additional heat treatment at 350°C, and Ti_3Ni_4 particles precipitate mainly on the subgrain boundaries. The temperatures of martensitic transformations and the deformation connected with RSME change in correspondence with a dissolving and precipitation processes of particles. Heat treatment of rolling samples at 500°C decreases transformation temperature on 30°C and the deformation of RSME in three times. Following heat treatment at 350°C increases transformation temperatures and the RSME deformation. Hence, special type of structure memory exists: the memory of RSME. Because of precipitating in aging Ti_3Ni_4 particles have a orientational connection with B2-matrix and martensite nucleates on the particles, so increase of the RSME deformation after aging at 350°C is explained by the reproduction of texture in particle arrangement. Thus we connect the memory of RSME with texture memory of Ti_3Ni_4 particles. The dissolving and precipitation process can be repeated. Heat treatment at 500°C and 350°C leads to or decrease of RSME deformation or increase, alternately. Note , that these particles play double role. The particles precipitation and the attendant impoverishment of matrix of nickel increase martensitic transformation temperature. The texture of particles and nucleation of martensite on particles



result in selection of martensite orientations and macroscopic deformation RSME. Therefore martensitic transformation temperature and deformation of RSME change synchronously. It is noted, that the processes of dissolving and precipitation of particles take place in the deformed and polygonized B2-matrix below recrystallisation temperature. That is the memory of austenitic type RSME realized [1-3].

One change of the deformation RSME and martensitic transformations temperatures for precipitation and dissolving of particles Ti_3Ni_4 observed for investigation of martensitic type RSME. The sing of martensitic type RSME is opposite of one austenitic type RSME. After deformation in 12% at room temperature and heat treatment at 500°C deformation martensitic type is near 0,15%. Deformation of RSME increased in 2.5 time and simultaneously increase temperatures of martensitic transformations on 25-30°C after additional aging at 350°C. The change of deformation of RSME and temperatures of martensitic transformations for precipitation and dissolving of particles copy multiply the same way as in case of austenitic type RSME. Obviously, it can be explained by reproduce texture in arrangement of particles, fixing the formed dislocation structure.

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CU-AL SHAPE MEMORY ALLOYS AND ACTUATORS

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The wide range of ternary and multicomponent poly- and single-crystal alloys based on Cu - Al - Ni and Cu - Al - Mn dispersion-hardened by ZrO₂ and HfO₂ particles has been developed. By transmission electron spectroscopy and small-angle X-ray has been established that the size of oxide fragments lays within the limits 30 - 100 nm. The shape memory effect and stress generation have been investigated at different conditions of mechanical and thermo-cycling under applied stress.

The principles of construction of actuators based on shape memory alloys in which the force element works in bending, tension or compression are developed.

The force element consists of shape memory alloy rod or plate, electrical heater and special system for fixing it into the actuator.

Tension or compression force elements can generate large forces, but the displacement will be relatively small. To obtain significant displacements the length of the force element should be noticeably increased, or otherwise a complimentary device for displacement transformation will be needed.

The bending force element creates significantly more displacement but still not as large as force than tension and compression force elements.

The structural schemes of linear and rotational cyclic shape memory actuators capable of generating force, moment and displacement in one or two directions have been developed.

In case of generated useful force and displacement in one direction one force element is used, while for the reverse movement steel spring is used.

In case of generating useful force and displacement in two directions two force elements are used.

An important technical characteristic of the actuators in question is the relation of the generated force (in case of the linear actuator) or of the generated moment (in case of the rotational actuator) to the dimensions and weight of the actuator itself.

The use of shape memory alloys makes it possible to create shape memory actuators that are able to generate large forces and moments, while the actuators themselves are by far smaller than the traditional, i.e. electromechanical, electromagnetic, hydraulic, pneumatic ones and other actuators.



However dimensions and weight of actuators essentially depend on properties of the shape memory alloys.

For advantage in dimensions the shape memory alloy should have the following properties:

- reversible strain not less than 6%;
- generated stress at least 150 MPa;
- no irreversible plastic strain under cyclic operation;
- stress required for the deformation of the force element should not exceed 30% of the value of the stress generated;
- the above properties should remain unaltered under cyclic operation.

Only in that case the smaller dimensions of shape memory actuators will turn out as an advantage in comparison with the dimensions of traditional actuators. This can be seen as an incentive not only for new applications, but also for substitution, in some cases, of traditional actuators.

Developed shape memory actuators have been used in drives and adaptive grippers of robotic systems for various applications.



INFLUENCE OF A TEMPERATURE CYCLING UNDER LOAD ON SHAPE MEMORY EFFECT IN CuZnAl SPIRAL SAMPLES

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Samples of the spiral shape represent major interest as a prototype of power elements of constructions that use the shape memory effect in their operation. In this connection the study of a behaviour stability of such elements was undertaken, the samples being cycled with transition over an interval of transformation temperatures.

The experiments were carried out on spirals manufactured of a wire by a diameter 3,5 mm of the alloy CuZn18Al7. A spiral was winded on after a wire annealing at 800 °C during one hour and quenching in water. Samples consisted of 12-14 coils with an external diameter of 18-20 mm. A mass of a sample was 36-38 g. Temperatures of the beginning A_s and the end A_f of inverse transformation varied for different samples: A_s - from 45 up to 55 °C, A_f - from 55 up to 65 °C . The rate of temperature changing was held at the level of 2-3 °C per minute.

It was reported in [1,2], that deformation behaviour of spiral samples during transition of temperatures of transformation is complicated. Simultaneously to change of the linear size of a spiral its twisting takes place. In a nonloaded condition this process is one-sided - the twisting goes on a direction of a spiral winding on. It is revealed, that the presence of a squeezing loading changes character of rotational process: it becomes two-sided. The primal rotation of a sample against a direction of a spiral winding on displays that the exposure promotes an untwisting of a spiral. The linear size at this stage practically did not change; the angle of rotation reached 10-15 degrees. Simultaneously with a kick-off of the height change of a spiral the direction of rotation was changed. The linear size of a sample increased by 140%, and the rotational displacement at the second stage reached 30-35 degrees. The temperature of the beginning of the linear size deformation changes depended on force, squeezing a spiral, - the more was the force the higher was the temperature when the deformation started. The deformation processes calling rotation of coils, on the contrary, began at one and the same temperature for different magnitudes of loading. The distinction in temperature intervals of two deformation processes, apparently, is caused by the external force precluding the linear size form changing. Simultaneously



there are internal stresses in a material of a spiral sample stimulating deformation changes in a direction, perpendicular to the external force, as soon as the transformation temperatures, typical for nonloaded metal, are achieved.

After the first thermocycles the nonloaded sample has not restored the primal shape after cooling [1,2]. Distance between coils became approximately 2 mm, that has increased the linear size of the sample from 52 mm up to 78 mm in the low-temperature state. This size did not change during 10 thermocycles of a spiral. The similar sample was subjected to a multiples temperature cycling with transition from one phase condition to another at presence of a compressive force 10 H. This loading ensured dense adjoining of coils of a spiral in a low-temperature condition, i.e. distance between coils was equal to zero. After 96 thermocycles the sample was liberated from a loading. In following thermocycle at heat the linear size of a spiral has achieved the peak magnitude in the same way, as at the previous heat, and after cooling returned the shape that was initial for the 97-th thermocycle. The clearances between coils arising in nonloaded samples after first thermocycles , missed. Thus, as a result of a temperature cycling on load the overgrowth of magnitude of a reversible deformation of a nonloaded spiral sample is obtained. However, in the following thermocycles it was registered that the rotation at the presence of a squeezing loading at heat has become to occur simultaneously to change of the linear size of a spiral and flowed past only in one direction - twisting. In our opinion, the changes of the deformation process character and the shape of a spiral after its temperature cycling on-load points to the internal stresses fields deviations in the material at a room temperature at the presence of a compressive force.

Total of the heat-cooling cycles sustained by a spiral sample for 1,5 years of trials, has made 200, that confirms the opportunity of engineering use of materials under consideration in multicycle settings.

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SHAPE MEMORY EFFECT IN Co-Ni ALLOYS

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The completeness at manifestation of a development of such properties as a shape memory effect, superelasticity, superplasticity and high damping connected with martensitic transformation depends on peculiarities of the martensitic transformation and thin structure (substructure) of martensite. In case of thermoelastic martensite transformation (MT) and definite type of a thin structure of martensite (twins or stacking faults) the full recovering is observed at heating (reverse MT). On transformation with large hysteresis and thin structure of martensite as dislocations and twins the initial shape is recovered partially.

Alloys Co - Ni (0-34 wt. %) with low energy of stacking faults are situated between the above mentioned groups of alloys. During MT the face-centered cubic lattice (α -phase) is reconstructed to hexagonal closed - packed one (ϵ' - phase). The transformation occurs with hysteresis, which depends on nickel content. In arising at cooling martensite the high density of stacking faults is observed.

Using deformation curves obtained by a method of four-point bending on cooling and heating the dependencies of the martensitic transformation parameters (characteristic temperature of the beginning and finishing of the direct and reverse transformations, intervals of the direct and reverse transformations, hysteresis of the transition) on Ni concentration and thermocycling in the transformation interval and also the coefficient of recovery of the initial shape have been investigated.

The microstructural changes in α - and ϵ' - phases in a quenched condition and after one, five and ten cycles in alloys Co - 31.5 wt. % Ni and Co - 20 wt. % Ni were studied using electron microscopy. The relation of microhardness to concentration of nickel was also established.

Increasing of Ni content from 0 up to 30 wt. % resulted in decreasing temperatures of a beginning of the direct MT from ~ 400 to 5 °C, reverse MT from 430 increasing to 150 °C, and of the hysteresis value from 50 to 200 °C. The coefficient of recovering of initial shape depends little on the Ni contents and are equal to ~ 25-40 % (fig. 1). On thermocycling the minor alteration of change of the characteristic transformation temperatures is observed. The influence of cycling on recovering has definite features: during the first 3-4 cycles the coefficient of recovering of shape is increased and then stabilized remaining practically invariable. In fig. 2, as an example, the relation of coefficient of shape recovering



to number of cycles for alloy Co -15 wt.% Ni is shown. Such behavior of shape recovering coefficient is may be caused by accumulation in the initial and martensitic phases dislocations that results in alloy strengthening (phase strain hardening). The electron microscopy has confirmed these supposition. On increasing of thermocycling number the dislocation density in both phases (parent and martensitic) was considerably increased.

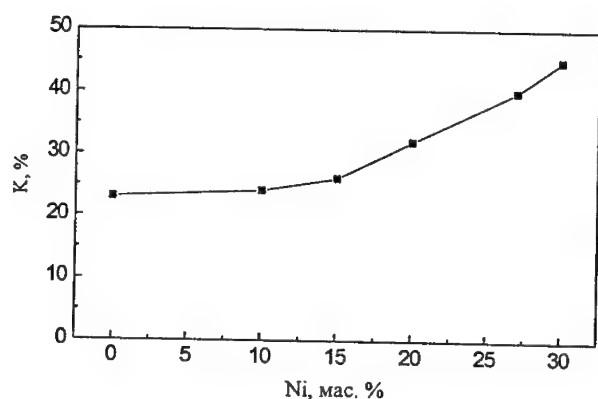


Fig.1. Relation of the coefficient of initial shape recovering on the Ni contents

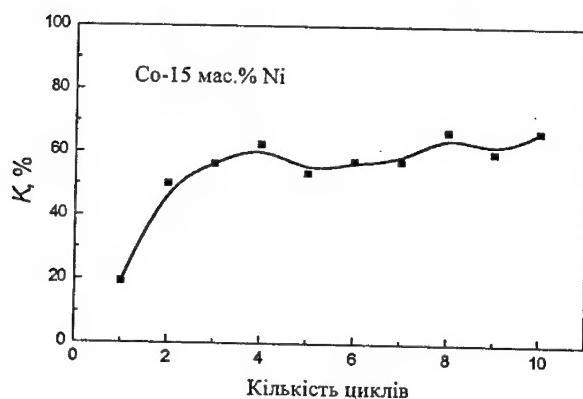


Fig.2. Relation of the coefficient of shape recovering to number of cycles for Co -15 wt.% Ni alloy



THE FUNCTIONAL MATERIALS IN PASSIVE SAFETY MEANS

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The safety of a power plant (PP) designed for any application can be ensured by: • defense-in-depth protection; • highly developed inherent safety of PP; • use of the proven engineering solutions; • capability to fulfil the assigned functions in case of failures; • safety systems (SS) (for shutdown, holding PP, heat removal, etc.) • based on the passive principle of action. In its functioning the passive safety mean (PSM) is triggered only by an event that caused the system actuation and does not depend on functioning of the active system (for control, power supply and other purposes).

PSM: • recognizes the value of required process parameter; • generates a scram signal; • amplifies and converses this signal; • switches over the retaining-starting device; • performs the protective action by moving the working organ of the actuator.

PSM is provided with the means enabling its functions (for heating and/or cooling, removal of gases, etc.). Responsible PSMs transfer the data on status ("available", "came into action", etc.). PSM is returned to its initial state (rendered) only by the operator.

Potential demands in and possibilities for designing of PSMs are especially appreciable in the any areas:

- based on process parameters: temperature, pressure, flow rate (its rise and backflow), level, neutron flux, etc.;
- based on the principle of the scram signal generation: heating, change in pressure drop, expansion, compression, buckling, poor heat removal, change in current, etc.;
- based on the principle of action: increase of resistance (hydraulic and electrical), change in balance of forces, melting, damage, rise of thermal emission, transition of ferromagnetic to paramagnetic, transformation of martensite to austenite (the effect of shape memory), superelasticity, etc.;
- based on the acting forces (impacts): mechanical constraint, gravity force and Archimedian force, electromagnetic and hydrodynamic forces, structural transformation force, etc.



PSMs may include different functional materials (i.e. those intended to perform functions) such as metal glasses, melts with changeable electric conductance, capillary (thermal) tubes, detonating fuses, gas-generating, gas-liquid, ceramic-metal, variable color coatings, lyophobic (non-wetting) capillary, with the shape memory effect.

Shape memory alloys (SMA) are among the most advanced and promising materials.

At present PPs are equipped with the following range of PSMs applications:

- safety devices with compressed springs; • retaining devices with magnets;
- rupture disks and membranes; • safety fuses; • thermal-mechanical weld reinforcing elements based on SMA; • compact valves with SM-actuators;
- repair valves with easy-to-remove SM-actuators which are stored in the warehouse and mounted for repairing the pipelines, if necessary; • ceramic-metal membranes (yes-gas, no – liquid); • hydrophobic platinum hydrogen neutralizers;
- fire-extinguishing automatic SM devices for penetration closure,etc.

It is envisaged reasonable in this century that the control logic involving skip of the first scram signal and waiting confirmation of its validity which is well-adopted in the electronic systems be also extended to PSMs. The following devices shall be applicable as:

- direct-acting devices controlled by the working fluid following the logic “1 of 1” and, if possible, “2 of 3”;
- passive controllers with their components being connected into the control circuit to ensure the actuating logic “2 of 3”;
- fast-acting passive initiators of action which respond directly to change in the process parameters and passive initiators which break the controller power circuit by the logic “2 of 3”.

The strategy for the use of PSMs implies the following: • introducing of the passive elements of normal operation, • addition of the passive initiators to the active PSMs, • systematic introduction of well-developed PSMs.

The following devices are available for application:

- direct-acting retaining and actuating devices of in the modular and easy-to-replace design; • passive initiators of actuation responding to loss of power, pipe ruptures, rise of ambient temperature and pressure (in the compartments), drop of level, increase of neutron flux, flow rate increase and backflow, • fuses, • indicators based on SMA, etc.



**INFLUENCE OF THE SYMMETRICAL TWINING BOUNDARIES ON
THE STRENGTH OF PROPERTIES OF HCP ALLOYS**

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Alloys with effect of shape memory are used as constructional, electro-technical, precisional, reactor materials in aircraft, medicine and other spheres of science. It needs the application of materials, characterized by its ability to restore the definite form at the definite external conditions. It is well known [1], that physical and physico-mechanical properties of alloys, for example: strength, elasticity, are determined by the presence of different defects in them.

In this paper it was studied the influence of the twinning symmetrical boundaries on strength characteristics of alloys Ti_3Al (superstructure $D0_{19}$) and Ni_3Ti (superstructure $D0_{24}$) by the method of computer simulation. It was carried out the calculation, using the pair interatomic potential, taking into account the anisotropy of superstructures on the basis of hcp lattice. The obtained values of twin energies formation were found in the interval from 100 mJ/m^2 to 400 mJ/m^2 . These values were corresponded to the experimentally observed ones [2].

Fig. 1 shows the patterns of atomic displacements near twinning planes $(10\bar{1}2)$ and $(11\bar{2}2)$ relatively to the hypothetical start configuration in alloy Ti_3Al . They are corresponded to the symmetry of the twin: atoms, situated on twinning plane, do not have normal displacements. Atoms, located on the planes, which are situated near the twinning plane, displace from boundary, forming local tension zones of the lattice. These zones are advantageous for distribution of interstitial atoms. Besides, splitting of the nearest twin planes into subplanes is observed. Aluminum atoms of defect crystal part are repulsed from local compression zones, having maximum displacements. Area of the relaxation displacements occupies the first 5 or 6 planes of each side from the twin.

Application of sphere symmetrical potentials led to a similar deformation of the crystal, having the largest amplitude of atom displacements. In this case values of twin energy formation appeared to be equal $\sim 1 \text{ J/m}^2$. These values exceeded well known experimental data.

¹Hirt J., Lotte I. Theory of dislocations. – M.: Atomizdat. 1972.-600c.

²Feng C.R., Michel D.J., Crowe C.R. A shot note on the microstructure of $TiAl/Ti_3Al$. // Scripta Metallurgica et. Materialia. – Vol. 24. – pp. 239-240. – 1990.



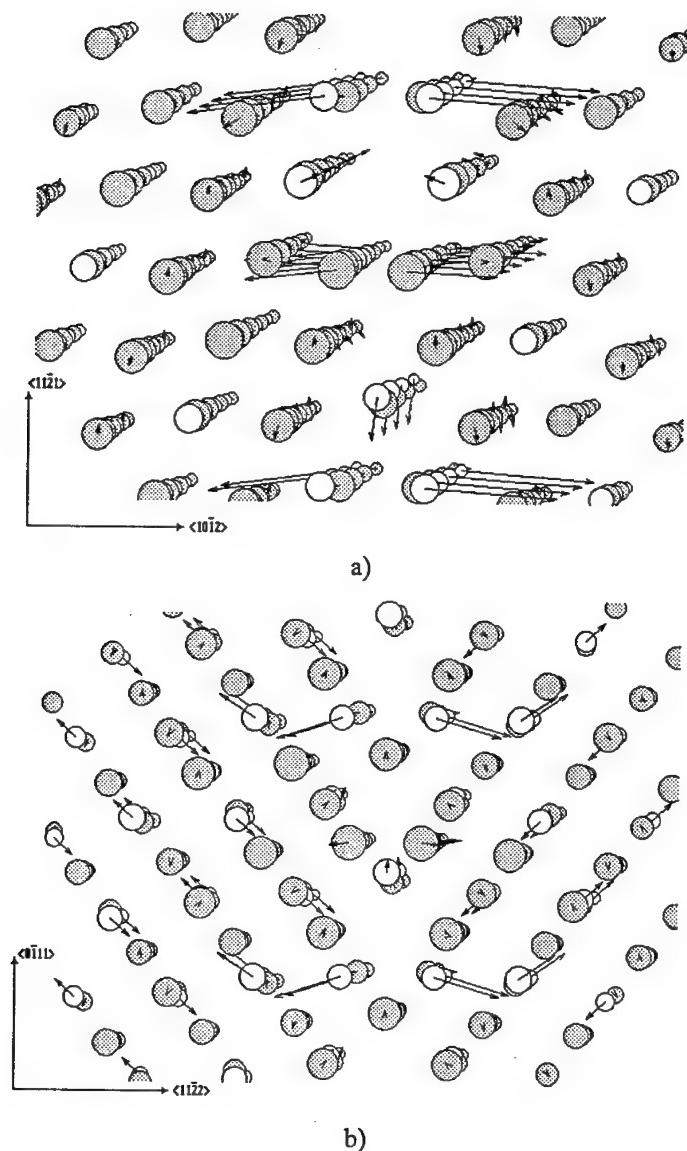


Fig. Patterns of atom configurations and microdeformations of Ti_3Al alloy lattice near the twins projection on the plane: a) $(1\bar{1}00)$, b) $(\bar{1}100)$. The scale of atomic displacements: a) 8:1, b) 4:1. White circles – atoms of Al, grey circles – atoms of Ti.



THE THERMOCYCLIC DURABILITY OF SHAPE MEMORY THERMOSENSITIVE ELEMENTS

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Thermorelay is the most representative type of martensite microdrive. It is meant for deflection of an elastic plate (of a contact holder) at 0,4-0,7 mm, which is enough for making sharp commutating contacts unlocking or bridging. A Thermosensitive element made of the shape memory effect alloy, is the force organ in such a relay.

At the present, the considerable enough number of thermorelays of different functions is already made, including [1-4], however, the items remain unstudied that are connected with their durability during the long time ($\geq 10^5$ thermocycles).

In this paper there are given the results of investigation of thermocyclic durability of thermoelements in a form of cantilever beam in conditions of the final binding rigidity. The St60S2 elastic spring was the counteracting body on the stage of generation of reactive forces of thermosensitive element. Thermoelements were made of the Ni-Ti alloy (50,5 at.% Ni). Investigation of thermocyclic durability was done in the interval of incomplete martensite transformation (MT) (30-50°C) [5]. The thermoelement bending deflection in condition of the counteracting spring was 0,52 mm, stress under heating till the operation temperature ($T_{op}=50^\circ\text{C}$) developed till 0,14 kg.

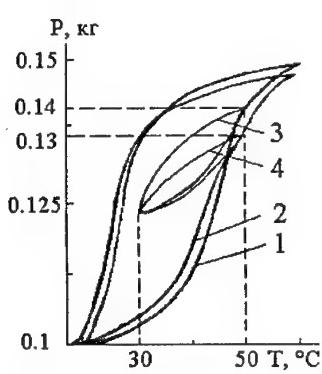


Fig. 1. The hysteresis loops in complete (1, 2) and incomplete MT intervals (3, 4); 1, 3 — loops after 1854 cycles; 2, 4 — loops after 6509 cycles

On Fig. 1 there are shown the MT hystereses in the MT complete and incomplete intervals of the Ni-Ti alloy (50,5 at.%). The small hysteresis loops are formed in the given interval of temperatures $50 \leftrightarrow 30^\circ\text{C}$ in condition of counteracting of counterspring with rigidity 0,05 kg/mm. It is seen that the given level of reactive forces in the incomplete MT interval, to the extent of increasing number of cycles, fall till the level of 0,13 kg after 6509 cycles, and till 0,12 kg, at the 10000th thermocycle. It is seemingly determined by relaxation of reactive stresses on the stage of cooling, due to the effect of

elasticity of direct transformation. The bending deflection, at this, has undergone little change: from $f_0 = 0,52$ mm till $f = 0,48$ mm after 10^4 cycles.

However, one should pay attention to the fact that thermocycling of the even unbinding sample always causes shift of temperatures of the return martensite transformation (A_{in} , A_f) to a greater degree than M_f and M_{in} . In this connection one can expect change of all the operation parameters (P_{max}^{in} , $\Delta\lambda$, T_{med} , T_{min} , etc.) at thermal interchanges during more than 10^4 cycles.

Thus, the investigations made for the first time, of durability of thermoelements from the shape memory alloy on the base of 10^4 thermocycles, showed some 8-10% fall of the level of the given force, as well as change of the bending deflection and duration of cycle in connection with accumulation of damage in the material, form-changing of the hysteresis loop and relaxation of the reactive forces under thermal interchanges.

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CONTROL FORM RESTORATION OF SME-ALLOYS WITH ULTRASONIC VIBRATIONS

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In [1, 2] the possibility of ultrasonic stimulation of the shape memory effect (SME) is shown. The exclusiveness of this method is that it allows to have control over the deformation process at the expense of the local warming-up of a material. The warming-up degree of the specific part of the material is the result of dissipation of the elastic ultrasonic vibrations energy in it and is determined both by the parameters of ultrasonic vibrations and dimensions and geometry of the sample. In its turn, strong dependence of damping properties and heat capacity on the change of the phase composition determines the evolution of temperature fields. It allows to set a complex program of work for the construction elements with the SME, successively influencing them by ultrasound.

The mathematical modeling of heat emission in samples of resonance length was carried out. Mathematical formalizations of the model to lead a mixed edge task for the non-linear equation of heat conductivity [3]. In this case the heat conductivity equation is single-measured:

$$\frac{\partial T}{\partial t} = \frac{\partial}{\partial x} \left(a(T) \frac{\partial T}{\partial x} \right) + \frac{a(T)}{\lambda} \varphi(x, t),$$

where T - temperature, $a = \frac{\lambda}{c\rho}$ - heat conductivity coefficient, λ - heat

conductivity, c - specific heat, ρ - material density, $\varphi = \frac{\delta \cdot U(x)}{\tau}$, δ - coefficient of elastic energy dissipation, U - elastic deformation energy, τ - period of vibration. The rheological part of the mathematical model consists of the description of the heat conductivity coefficient behavior and $\varphi(x, t)$.

The results of experiments of locations (fig.1) and those of numeral ones (fig.2) on the ultrasonic warming-up of samples of resonance length TiNi and CuAlNi showed their good coincidence.



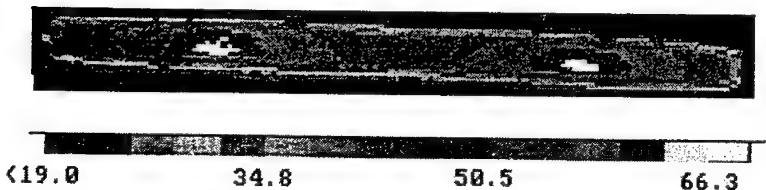


Fig. 1 Thermovision temperature profile of the TiNi sample of wave length.

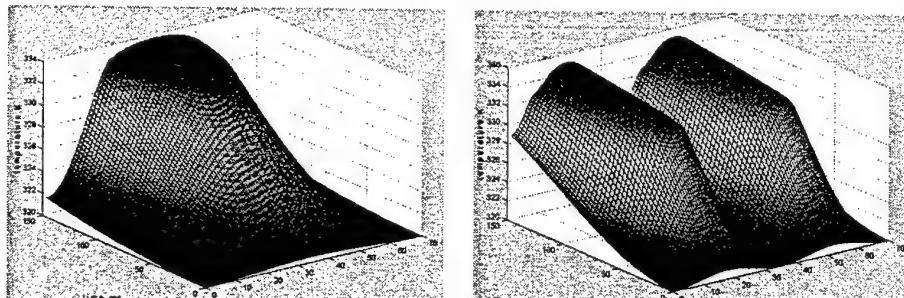


Fig. 2. Evolution of TiNi temperature field at ultrasonic heating for a resonance sample of half-wave length (on the left) and that of wave length (on the right).

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SINTERED MATERIALS

189-210



DEFORMATION OF POWDER BODIES OF SYSTEM TITAN- NICKEL AT SINTERING

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Proposed by Professor Savitskii A.P. model of the two-component powder body allows not only to explain conformities at sintering in solid phase, but and predicting conformities of deformations of powder bodies at liquid phase sintering. In the capacity of elementary volume of two-component powder body indicates the group coming into contact to each other particles of arbitrary form, the central one of them is component – addition, and others which surrounding in the one layer – the component, which is the basis of powder body, and free spaces between particles – pores between particles.

In this work of PTOM- titan and PNK- nickel powder were used with dispersity of particles lower than 50 microns. Reached by method of pressing samples was sintered in the interval of temperatures 650-1000°C with isothermal endurance while one hour.

According to primary direction of diffusion flow of atoms from one component to other, prove to be by them nature altogether definite and constant for given binary system, in the sintering process with reference to each other one of the components, as a rule, is a activator, and other is a deactivator. Thus, in system the sign of deformation of powder body is defined with that, which of components is forms the basis of that body, and which is an additive.

Comparison of deformation character of the samples in titan- nickel system in frameworks of signed higher model showed, that partial coefficient of geterodiffusion of nickel to titan. Since that primary direction of diffusion flow of titan atoms into nickel particles is definite in whole temperature interval of solid phase sintering, consequently, that at solid phase sintering titan is a component primarily transferring in diffusion mechanism.

At the transfer from solid phase sintering in conditions of liquid phase sintering process of alloy forming accelerates so far as contact square of diffusion interaction of components essentially increases in consequence of spreading of alloy to particles of solid phase under action of capillary forces.

Analysis of concentration dependence of samples deformation at transfer from solid phase sintering to liquid phase sintering shows, that appearance of liquid phase in sufficient amount take place at 850°C, that is approximately on 100°C lower of melting point of low temperature eutectic, that connected with self heating of samples in consequence of heat yielding of intermetalloys forming. Since at liquid phase sintering character and sign of deformation of samples “basis-additive”, like in a case of solid phase sint5ering, stay constant, that primarily



direction of diffusion flow of titan atoms into nickel particles becomes definite and constant characteristics of given binary system, independently from sintering kind. We must note to one circumstance, which have a importance in establishment of conformities of samples deformation of binary systems at solid phase sintering. With transition from solid phase sintering to liquid phase sintering displacement of shrinkage and growth maximums took place to side of increasing of additive maintenance. Given location in condition diagram reflects line of changing of top significance of limited solubility of components in solid condition. At this process of shrinkage breaking, which is seen at solid phase sintering, in presence liquid phase is replacing by samples growth with titan additive. Given samples growth process with titan additive continue also at further increasing of sintering temperature.



DEFORMATION HARDENING OF ALLOYED STEELS

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In many cases the operation of powder blanks' deformation are indispensable for maintenance of demanded structural and properties of an article. During deformation at various temperature the dynamic hardening develops to some extent, the knowledge of a kinetics which one will allow to control grain structure of a powder material.

Deformation hardening of powder materials is accompanied by two simultaneously developing processes: by change of structural porosity and hardening of a solid phase. Connection between parameters of a structure and characteristics of a intense condition during deformation, and especially, influence of compressing stress on change of porosity and on structure of solid phase are not established for many materials and, in particular, for steel produced of a dust "Distaloy". That is why the analysis of deformation hardening at axial compression of powder materials on this basis is interested.

The purpose of this article is the analysis of change of porosity and deformation hardening of powder environment at an axial compression.

The researches designed on samples produced of a dust "Distaloy" AB+0,8% - wax of a following chemical structure: 4,0% Ni, 0,01% C, 1,5% Cu, 1% Mo. Cold two-sided pressing of exemplar by the size 8×16 mm are executed on a testing machine ZD-4 Then this samples are sintered at temperature 1100°C in environment of hydrogen. Porosity was varied in a range 8-30%. Further samples are tested for compression on a testing machine ZD-4 with an automatic record of the diagrams of compression. The accuracy of fixation of load value made not less than 0,5%, the deformation are not lower 0,05%. A strain rate was 10^{-3}s^{-1} .

The curves of hardening in coordinates $\sigma - e^{1/2}$ are plotted, where "e" was a logarithmic degree of deformation.

They completely conformed to legitimacy of deformation hardening process of powder materials. Three sites of curves of hardening are obtained which one conformed to different stages of deforming. And the miscellaneous slope concerning an abscissa axis is characterized for each site.

The curves of solid phase hardening of samples with a different porosity concurred practically completely in all investigated range of porosity. The curve of a solid phase hardening also consists of three legibly expressed straight-line sites, describing different structural condition of a material at deforming.

It is possible to assume, that in this condition the processes of deformation occur, as well as in a compact material. For the first site the sharp increase of a pressure is observed as a result of hardening, that is connected to formation of congestion of dislocation, on the second site the intensity of hardening decreases, and



the conditions for formation of cellular structure are created which, apparently, is formed at the third stage.

The meanings, designed from experiment, of factors of hardening normalized on the module of shift $K = k/G$, depending on a degree of deformation of sites on a curve of hardening are submitted in tab.

Table
Factors of hardening at three stages of deforming

Den-sity, θ, %	1 stage		2 stage		3 stage	
	σ_0 , MPa	$K \cdot 10^{-6}$	σ_0 , MPa	$K \cdot 10^{-6}$	σ_0 , MPa	$K \cdot 10^{-6}$
29,5	24,8	7,5	63,1	3,59	86,2	2,43
15,2	36,2	10,5	128,6	3,89	150,1	2,66
10,6	39,5	12,1	138,7	5,05	163,5	2,78
9,6	48,5	13,9	147,1	5,31	175,2	2,80
8,7	50,01	14,9	148,5	5,56	180,0	2,89

It is necessary to note, that with increase of a degree of deformation the stress intensity σ_0 at each stage grows, that testifies to growth of hardening at the expense of change of a dislocation constitution of a solid phase and reduction of volume of pores.

Factor K decreases with increase of a degree of deformation and at last stage its size reaches meaning close to meaning of a solid phase. Normalized on the module of elasticity the factors of hardening of a firm phase make: 1 stage – $13,4 \times 10^{-6}$; 2 stage – $6,65 \times 10^{-6}$; 3 stage – $2,42 \times 10^{-6}$. It testifies to uniform enough distribution of deformation at plastic flow of an investigated material owing to high homogeneity on a chemical structure of the given powder alloy.

At an axial compression of steel produced of a dust "Distaloy", two connected processes are observed: modify of solid phase structure and change of porous structure. The estimation of hardening is executed with use of initial stress which grows with each stage of deforming, and factor of hardening describing intensity of hardening process. At a degree of deformation 18-22 % and normalized factor of hardening less than $2,89 \times 10^{-6}$ the process of hardening is executed basically for the account of a solid phase deforming.



RESEARCH OF TECHNOLOGICAL PROCESSING OF COPPER SCRAP BY POWDER METALLURGY METHODS

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About 40 % of copper scrap is a current conductors, executed from copper with purity not less M1 (99,9% Cu) and polluted by isolation. The main way of processing of such scrap is melting repartition with subsequent electrolytic refining and melting of cathodes. This method of processing requires significant power and labour costs.

Main laws of processing of copper scrap to finished products or rod blanks by powder metallurgy methods are investigated and established.

In a initial condition scrap is a chaotic twisted strands of the various form and length. For removal of isolation and transfer of copper to a fragile condition, that provides the intensification of reducing to fragment process, scrap was annealed in hydrogenous environment. Therefore copper becomes fragile, and isolation become charred. Some features of development of hydrogenous illness of copper is established by the tests to bend on ГОСТ 1579-93 and metallographical analyse and optimum rates of annealing is determined. After processing in sphere mill and aeroseparation copper with purity not less 99,9 % is produced in a kind of fibres by a length of 1-20 mm.

Depending on a diameter and the lengths fibre have bulk density from $2,5 \times 10^3$ up to $4,6 \times 10^3$ kg / m³ and they fill in a cavity of the press – form very well, as in hand-operated, as automatic rates. And, than less attitude of a length to a diameter of a fibre, themes more bulk density.

Process of condensation of fibres the diameter from 0,6 до 1,3 mm and various length is investigated. It is established, that at a initial stage the condensation is defined by bulk density of fibres, and after density of sample is of the order $7,3 \times 10^3$ - $7,5 \times 10^3$ kg / m³ the character of condensation for fibres of the various sizes is identical. The maximum density at pressure 780 MPa has made $8,85 \times 10^3$ kg / m³. After pushing out of sample from the press - form elastic sequel is occurred and the density is reduced up to $8,650 \times 10^3$ - $8,70 \times 10^3$ kg / m³. The size of elastic sequel depending on the size of fibres and pressure of condensation makes 2,0 - 3,0 % on height and 1,0 - 1,5 % on a diameter.

As it was established by measures of fibres of destroyed samples a degree of deformation of a solid phase is fluctuated from 15 up to 45 % and it is increased at condensation of fibres with smaller bulk density, that is agreed with literary data and is confirmed by account under the formula offered by Dorofeev U.G. for definition of a degree of deformation of a metal at condensation of powder blanks.



For removal of riveting after condensation samples are annealed in environment of dissociating ammonia in a range of temperatures from 550 up to 950 °C. It is established that as a result of relaxation of accumulated stress during condensation, growths of samples are occurred, increased with augmenting of pressure and temperature. Only after eight-hour endurance at temperature 950 °C the further increase of the sizes are ceased.

In annealing process in fibre material a number of parallel proceeding processes, stipulated by sizes of fibres, by their structure and density of sample is observed. As a result of development of recrystallization processes a structure of copper is changed, processes of consolidation are developed, though the density of samples is thus reduced.

For reception of density of equal density of castings and deformed copper and high property researches the processes of hot stamping fibres blanks are conducted. For this purpose the blanks by density $8,55 \times 10^3 - 8,60 \times 10^3 \text{ kg / m}^3$ heated up in protective environment on temperature 820 °C and stamped in a closed stamp on a hydraulic press by pressure 780 MPa. Received samples had residual porosity 1,2 - 1,4 %. However, as tests on a upsetting and metallographical researches have shown, splicing of fibres occur not completely, and received samples have not enough high properties.

For maintenance of high density and creation of optimum conditions of fibres consolidation the circuit of stamping was used, at which significant shift deformations are developed. Blanks by a diameter of 18 mm are heated up in protective environment up to 820 °C and are stamped in a die by a diameter of 25,0 mm, and then repeatedly heated up and stamped in a die of 34,0 mm. The degree of deformation on the first transition of hot stamping made 36 %, and on second - 47 %. After annealing in protective environment at temperature 550 °C samples had density $8,88 \times 10^3 \text{ kg / m}^3$ and hardness 54 HB, that meets to hardness deformed and annealed copper. Microstructure of samples is small-grained. The size of a grain is 5 - 10 Mk.

Thus, technology of processing of a copper scrap, enabling on the basis of powder metallurgy operations and thermal processing to receive rob blanks of copper with density $8,88 \times 10^3 \text{ kg / m}^3$ is developed, which can be used for reception of finished products by stamping methods or machining.



STRUCTURAL TRANSFORMATION IN POWDER METAL RADIAL BEARINGS.

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 Alexander Shatsov (Scientific Center for PM, Perm)

Research in the field of powder steels parts running in ESP friction units allows to determine friction processes pattern in operation (conditions), enabling wear resistance and quality improvement.

ESP stage components were tested (for radial wear): radial sleeve* made out of ЖГр1Д3 material and diffuser bore made out of ЖГр1Х2МД15 material, based on metal powder of SC.100.26 grade and ПДКР 3.200.28 grade consequently with colloid carbon of ГК—2 grade. The following powders were used as alloying components: copper powder of ПМС—1 grade, chrome powder of ПХ1С grade and molybdenum powder of ПМЧ grade. The mixtures were compressed at 420 MPa and sintered in the furnace (СЛОН) at $t=900-1160^{\circ}\text{C}$. Diffuser bore samples sintering was effected alongside copper base infiltration impregnation.

The samples exposed to abrasive wear in water conditions were tested. The tests were run on a specially designed test loop (stand) that permits testing of 5 stages in field-like conditions. Abrasive particles content (SiO_2) amounted to 0.2% by volume of the fluid (2gr/litre). Average particle size was 15 micron (size range 2-100 micron).

Radial sleeve sample has a characteristic pearlite structure. Diffuser bore sample has a martensitic structure with a portion of residual austenite of 10—15%.

As a result of research, fine structure change pattern in abrasive wear conditions in water medium was established, a correlation between intensity and gravity (density) of line defects of ρ lattice $I_h=f(\rho)$. A correlation between phase transformations in diffuser bore $\gamma \rightarrow \alpha'$ and gravity of line defects ρ was established. Wear rate variation I_h dependency on α' -phase quantity is established.

Research showed that a substantial decrease (approximately up to 2 times) in structural member (block) sizes occurs during the initial period of test, i.e. during the first minutes of it. Duration of tests has practically no effect on block size change.

Maximum microdistortions of radial sleeve sample lattice amount to 0.1%, minimum to 0.07%, and maximum microdistortions of diffuser bore sample lattice amount to 0.4% and 0.2% accordingly.

Line defect gravity calculation was done by diffractioal maximum widening.

With the test conditions being identical, line defect gravity is nonmonotonically dependent on test duration. $I_h(\rho)$ function is of cyclical character. Minimal value of wear rate for radial sleeve corresponds to line defect gravity $\rho = 1,5 \cdot 10^{11}$



sm^{-2} . These results confirm Yudin's hypothesis that with certain line defect gravity maximum frictional energy gravity (density) is obtained with corresponding minimal wear.

$I_h(\tau)$ and $\rho(\tau)$ relations have nonmonotonic cyclic character. Curve data maximum occurs during the first hour of test. During this period bedding-in of mating surfaces occurs with an increasing number of surface defects and wear rate intensity.

Quantitative analysis of change of phase $\gamma \rightarrow \alpha'$ showed that the proportion of α' -phase obtained amounts to approximately 4-7%.

*radial sleeve — a sleeve intended to protect a shaft from radial wear. It is inserted into the diffuser bore.



CALCULATION OF THIN-WALL CONSTRUCTIONS FROM SINTERED MATERIALS

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The perfecting of technological receptions of powder metallurgy in creation of original hardware products of new engineering (for example, filtering thin shells and diaphragms), it is impossible without correct application of analytical methods of calculations of strength properties of these materials and prognosis of their behaviour at extreme loads - rise of intrinsic pressure, mechanical and thermal shocks. It predetermines necessity of more careful and fluidized approach to design calculations of such constructions.

The matching of outcomes of numerical analysis of the equations of thin shells with the data of precision physical experiments displays considerable discrepancy of a rating and observable value of critical (destroying) pressure. For elimination of these mismatches we shall try more strictly and sequentially to take into account a physical nature and status of a material with reference to cylindrical thin shells of a construction from porous powder sintered material.

Let k_x, k_y main curvatures of a median surface in a point (x, y) , w - normal moving, Φ - stress function. Then for base equilibrium equations and the compatibilities of deformations in geometrically nonlinear production can be selected on [1] and are supplemented by conditions which are taking into account geometrical irregularities sintered material, for example, exposed so-called open it is time, which one change

parameters k_x, k_y . Let's enter the function $w_0(x, y)$ specifying of initial "abnormality" of the shell and we shall enter the parameter of Poincare - Lyapunov ξ [2], we shall designate amplitude of geometrical irregularities through. Is decomposable (w, Φ) , on three numerical parameters

$$w(x, y, \xi, \tau, \delta) \sim w(x, y, 0, 0, 0) + \xi w_{100} + \tau w_{010} + \delta w_{001} + \xi^2 w_{200} + \dots$$

$$\Phi(x, y, \xi, \tau, \delta) \sim \Phi(x, y, 0, 0, 0) + \xi \Phi_{100} + \delta \Phi_{001} + \xi^2 \Phi_{002} + \dots \quad (1)$$

By the known procedure we shall reduce a boundary value problem to recurrent of sequence. As a first approximation on ξ shall receive the task on eigenvalues for the parameter P . After the stipulated conversions we can write:

$$\frac{D}{h} \Delta^2 (w - w_0) = k_y \frac{\partial^2 \Phi}{\partial x^2} + k_x \frac{\partial^2 \Phi}{\partial y^2} + L(w, \Phi) + \frac{P}{h} \quad (2)$$



$$\frac{\Delta^2 \Phi}{E} = -\frac{1}{2} [L(w, w) - L(w_0, w) - k_y \frac{\partial^2 (w - w_0)}{\partial x^2} - k_x \frac{\partial^2 (w - w_0)}{\partial y^2}]$$

Here h - width of the shell, E - elastic modulus, P - pressure
 L - bilinear differential expression, Δ^2 - biharmonic operator Laplace, D - flexural rigidity:

$$D = \frac{Eh^3}{12(1-\nu^2)}, \quad (3)$$

ν - Poisson's ratio, (x, y) - Cartesian axials.

Let's consider the task about a rating of residual stresses under the explained circuit(scheme) in view of offered conversions. Let x_0, y_0 , about residual stress, and about (x, y) appropriate function, which one obeys (4):

$$\sigma_x^0 = \frac{\partial^2 \Phi^0}{\partial y^2}; \quad \sigma_y^0 = \frac{\partial^2 \Phi^0}{\partial x^2}; \quad \tau^0 = \frac{\partial^2 \Phi^0}{\partial y \partial x}; \quad (4)$$

To research the system (1) it is possible under the earlier explained circuit [2]. However information (experimental outcomes) about their allocation in the shell is indispensable for an actual rating of effect of residual stresses. Unfortunately she(it) frequently misses in substantial conditions of production of experiment. Are generated only technology of manufacture is model materials with rather small values about $\sigma_x^0, \sigma_y^0, \tau^0$. And in this sense the technology of a powder metallurgy creating materials, the technological porosity which one promotes a considerable stress relief of distorted materials, farly from being has reached the possibilities, on as indicate calculations on tendered adapted technique.

Calculation on the offered equations (2) minimum widths (h) in the shell of construction from sintered material representing high porosity filtering section as the cylinder $\varnothing 60 \times (50 \dots 80)$ mm, indicate possibility of saving of stability by such construction at rise of a working pressure (or operation - at contamination of pore channels before the next regeneration) up to 50 % without even of corrupting at wall thickness not less than 3,0 mm for hardware products on the basis of ceramic powders (Al_2O_3); 2,0 mm - on the basis of metallized ceramic powders (5-th % «pointwise» - shells" copper coating by width 4-6 microns on Al_2O_3); 1,0 mm - on the basis of powders Ferri lactas and stainless steels and not less than 0,5 mm - on the basis of powders of copper, brass and bronze.

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PECULIARITIES OF THE STRAIN HARDENING OF POLYCRYSTAL STRUCTURE IN SINTERING OF CUBIC BORON NITRIDE POWDERS

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The possibility for strain hardening of some covalent crystals having the diamond-type lattice, the correlation between dislocation structure characteristics and the crystal hardness have been considered in [1] and the relation between fracture toughness and dislocation density in as-synthesized cBN single crystals has been discussed in [2]. Our studies of the real crystalline structure of cBN polycrystals produced by sintering of cBN powders in a wide range of p,T-parameters have shown that there is a possibility to inherit defects of the crystalline lattice of the initial powders and to purposely form a deformation substructure in the course of sintering [3].

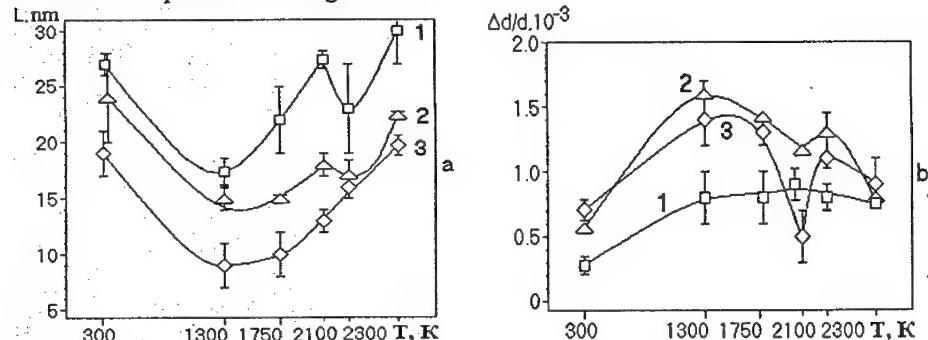
In the present paper we discuss the findings of our investigation into strain hardening of polycrystalline cubic boron nitride during the sintering of powders of various grit sizes (Table 1).

Table

Particle Sizes and Specific Surfaces of the Initial Powders

Powder types	KM1/0	KM2/1	KM3/2	KM5/3	KM7/5	KT-40	KM-40
D _{max} , μm	2.5	3	5	7	10	70	80
D _{med} , μm	0.6	1.4	2.3	3.5	4.9	29.2	38.1
D _{min} , μm	0.1	0.8	1.0	2	3	14	25
σ, m ² /g	12.2	4.54	2.75	1.81	1.27	0.223	0.131

Experiments have been carried out by the procedure described in [3] and the results are presented in Figs. 1 and 2.



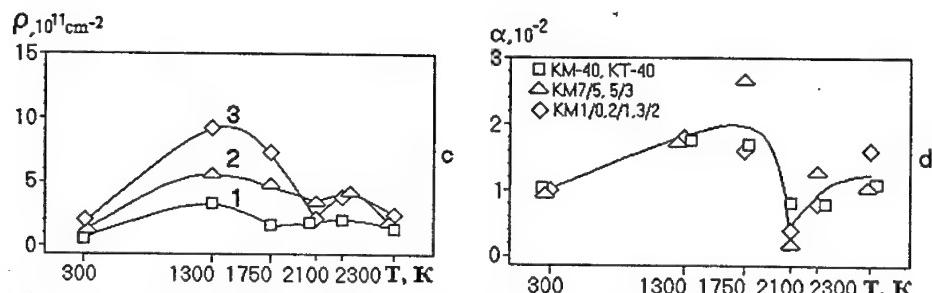


Fig. 1 Dependences of coherent scattering area (a), microdistortions (b), dislocation density (c) and stacking fault concentration (d) of the cBN crystal lattice on the temperature of sintering of polycrystals at 7.7 GPa. 1 - KT-40, KM-40; 2 - KM7/5, KM5/3; 3 - KM3/2, KM2/1, KM1/0.

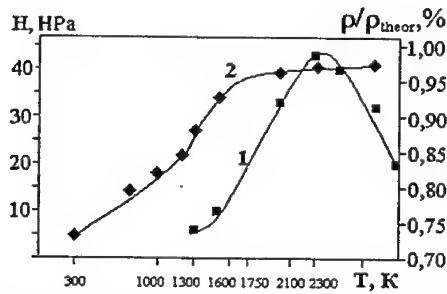


Fig. 2. Hardness (1) and density (2) of polycrystals vs. sintering temperature at 7.7 GPa, initial powder KM7/5..

A curve of the dependence of the density of cBN polycrystal structural defects on sintering temperature exhibits two maximums. The first maximum is in the 1300-1750 K temperature range and is related to the formation and growth of contacts. The fact manifests itself in the increased hardness and density of polycrystals. The second maximum is at 2300 K when the densification is practically completed, and the structure strain hardening corresponds to the highest hardness of polycrystals

The behaviors of the curves of temperature dependences of structural defects for powders having different initial grit sizes coincide. The quantitative characteristics indicate that sintered polycrystals inherit dislocation structures of the initial powders.

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CRACK GROWTH RESISTANCE OF Ti-Si INTERMETALLIC ALLOYS

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Novel intermetallic alloys are applied widely as structural materials for manufacturing of the components of the up-to-date high efficient power units (i.e. gas turbines, internal combustion engines, etc.), force drives (elements of screw propeller, blades, and augers which operate in aggressive environment), suspension elements of the magnetic pillow of trains etc. Among those, Ti-Si eutectic alloys alloyed by Al and Zr, possessing the high fracture toughness and corrosion resistance of titanium with high-temperature strength and hardness of ceramic structural compounds should be noted. This class of materials elaborated for the first time in Ukraine by materials science researches of the National Metallurgical Academy of Ukraine, has been developed further in metallurgical and structural aspects by scientists of this academy as well as the Francevych Institute of the Problems of Material Science of the National Academy of Sciences of Ukraine. Since the beginning of this research work – for more than 10 years - scientists of Karpenko Physico-Mechanical Institute conduct researches directed at improvement of the service characteristics of Ti-Si structural modifications carrying out the greater part of fracture mechanic tests.

The control mechanical tests of the primer modifications of Ti-Si cast composites (estimation of ultimate and yield strength as well as characteristics of plasticity and fracture toughness under static tensile and bending loading) in laboratory air in the temperature range of 20 ... 1000°C allow us to establish its mechanical stability margin. For example, at room temperature the ultimate tensile stress of 700 ... 800 MPa (at very low plasticity), ultimate bending stress of 800 ... 900 MPa, and fracture toughness of 10 ... 18 MPa m^{1/2} were established for these materials. The strength characteristics decreased two times at testing temperature of 700°C, and the relative elongation increased to 3 ... 4% while the fracture toughness did not change. The modification plasticity noticeably increased (the relative elongation was 13 ... 18%) and the failure stresses decreased abruptly (to 250 ... 300 MPa for both loading modes) while the testing temperature was elevated up to 800°C, and a material became insensitivity to stress concentration. The mechanical characteristics set obtained allowed us to recommend Ti-Si composites for manufacturing of important elements which operate under stresses in the temperature range of 600 ... 700°C.



The thorough tests of Ti-Si modifications under long-term static and cyclic loading showed a significantly higher creep resistance to compare to heat-resistant metallic alloys, and the satisfactory fatigue endurance in the temperature range mentioned above.

Because compositions of this class of materials are brittle at temperature up to 600 ... 650°C the static and cyclic fatigue crack growth resistance is a significant characteristic of service reliability of products made of these materials. The following structural and composition optimization of Ti-Si composites was carried out using just the crack growth resistance parameters. Fracture toughness of the novel hypoeutectic modifications is from 18 to 24 MPa m^{1/2} in the temperature range of 20 ... 600°C, increases up to 22 ... 26 MPa m^{1/2} at 700°C, and decreases sharply at higher temperature. Simultaneously fracture toughness of the hypereutectic modifications is lower (8 ... 14 MPa m^{1/2}) in the temperature range of 20 ... 500°C, but increases significantly with temperature growth, reaching 25 ... 28 MPa m^{1/2} at 700 ... 800°C. Respectively, such types of Ti-Si composites can be used in industry for different purposes.

Using results of the high-temperature cyclic fatigue testing of Ti-Si composites it was established that the fatigue crack growth rate diagrams (FCGRD) of the modifications are invariable in the temperature range from 20 to 500°C. While temperature rises up to 700°C the range of the threshold stress intensity factor ΔK_{th} decreases and the fatigue crack growth rate increases in Paris' region of FCGRD, however the cyclic fracture toughness ΔK_{fc} is changed insignificantly.

The influence of the service environment on the mechanical stability of Ti-Si composites was studied using a set of the electrochemical and corrosion-fatigue crack growth resistance investigations. The structural sensitivity of Ti-Si composites to hydrogen containing fluids as well as the dependence of threshold range ΔK_{th} on the phase composition of material were found.

Investigations on the mechanical and corrosion aspects of fatigue crack growth in Ti-Si compositions were carried out taking into account loading frequency, stress ratio, specimen sizes etc. Each of the noted factors has its own influence on the corrosion-mechanical behavior of specimens. Therefore, the laboratory simulation of the industrial service conditions of the products allows us to predict their serviceability or residual life with high accuracy.

The obtained scientific results, methods, accumulated data bank and experimental experience allow us to recommend Ti-Si intermetallic composites to be used by customers and also our services in evaluation of service reliability of industrial intermetallic products.



INTERCOUPLING OF PARAMETERS OF STRESS STATE WITH NATURE OF DEFORMING OF OSCULIFEROUSBAKED METAL

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Some articles, which are made on large experimental practice, and also own outcomes and the observation, indicate that influencing to a parameter index of rigidity of a state of stress "k" and parameter

μ_σ on process of deforming of an porous stuff close to bound with capacity to essential volumetric strains. As a result of development of deformations of a forming. And, in one conditions it there can be deformations of padding seal, and in other, to the contrary dilatation. To the given phenomenon apply the technical term "dilatation". But as in translation it means dilatation, that part of deformations of seal, that passes only at the expense of development of shearing stresses, call as negative dilatation or sometimes contraction.

At deforming is robust of all stuffs the bifurcation of curves $\sigma_i - \varepsilon_i$ i is watched also.

By decreasing of an initial porosity, probably to achieve the greater resistance to deforming and decreasing of a bifurcation of curves $\sigma_i - \varepsilon_i$.

For last of year by many scientists are made broad researches on definition of influencing to a parameter of a state of stress on a plasticity of metals constructed charts of a plasticity of oxide layers and welds studied relations of the charts of a plasticity to an elemental composition and doping of separate steels, from modes of heat treatment constructed charts of a plasticity at heightened temperatures, stored stuff about the charts of a plasticity of many steels, alloys and non-ferrous metals in a broad band of change to a parameter of a state of stress. Is necessary to mark, that for all investigated stuffs reference is increase of a plasticity with decreasing parameter «k».

Is established, that at a constant parameter of a state of stress parameter of a type of a state of stress has essential influencing on a plasticity. Sometimes at one «k», but different μ_σ of plasticity can differ in some times. The influencing to parameter μ_σ on plasticity is ambiguous, and depends on the physical nature of stuff and kept in repair addition agents. In this connection to neglect influencing to parameter ms it is not necessary, and the chart of a plasticity is recommended to be examined as boundary surface in space of two independent parameters of a state of stress «k» and « μ_σ » (under thermo-fast-track constant conditions of deformation).

For a clearing of the cause of flaking of curves $\sigma_i - \varepsilon_i$ there were investigated changes of volumetric strain in a function of intensity of deformations at loading on roads, which one are in deviators of a plane.



The known relation, with the help by which one established connection μ_σ with a angle of a type of stress state given as:

$$\mu_\sigma = \sqrt{3} \operatorname{ctg}(\omega_\sigma + \pi/3)$$

From last formula receives

$$\omega_\sigma = (\operatorname{arctg} \frac{\mu_\sigma}{\sqrt{3}}) - \frac{\pi}{3}$$

Therefore, the conforming functions $\sigma_i - \mu_\sigma$ are constructed in polar coordinate system at fixed values ε_i have nonmonotone nature. Allowing, also, minor bifurcation of curves $\sigma_i - \varepsilon_i$ at $\mu_\sigma = -1$ and $\mu_\sigma = +1$ bodily interquartile approximated relation of state:

$$\sigma_i = a + \epsilon * \cos\{6[\operatorname{arctg}(\frac{\mu_\sigma}{\sqrt{3}}) - \frac{\pi}{3}]\},$$

$$a = \frac{\sigma_i(\mu_\sigma = \pm 1) + \sigma_i(\mu_\sigma = 0)}{2};$$

Where

$$\epsilon = \frac{\sigma_i(\mu_\sigma = \pm 1) - \sigma_i(\mu_\sigma = 0)}{2}.$$



INFLUENCE OF TMT ON THE STRUCTURE AND MECHANICAL PROPERTIES OF TUNGSTEN RODS

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Tungsten produced by powder metallurgy technique is widely applied as a wire to manufacture details in lighting engineering and electronics. The wire is manufactured from a sintered tungsten billet of WA trade mark (0.006K, 0.006Si, 0.004Al, 0.005Fe, 0.03Mo in wt.%) with the porosity of 12-14% by the technique of thermo mechanical treatment (TMT)

Potassium is known to be practically undissolving in W, and the temperature of its boiling is 760C, therefore additives of K in the amount of 50-100ppm lead to the formation of bubbles filled with K. These bubbles situated in intercrystalline boundaries sharply increase the temperature of intense recrystallization.

The formation of the necessary complex of wire mechanical properties is performed in all stages of processing: billet – rod – wire.

The influence not only the most traditional method of deformation for manufacturing rod- rotation forging, but also new aspects TMT- a four-roll rolling or a screw rolling (more ecological pure and productive methods of plastic deformation at production of a tungsten rod) on the structure and mechanical properties of rods is investigated.

The sintered billet contains two types of pores: coarse pores about $1\mu\text{m}$ in size, and small pores about $0.1\mu\text{m}$ in size. In the processes of billet deformation and intermediate annealing coarse pores vanish, and small pores remain in all processing stages. They are well revealed when fracture is of intercrystalline character (fig.1, 2).

During warm plastic deformation of the sintered billet and intermediate annealings its compacting, healing the defect of crystalline structure as well as evaporation of potassium and formation of potassium bubbles in grain boundaries take place. The distribution of K in the body of W wire, the dimension of K bubbles and their growth in annealing process make an essential influence on the structure and properties of rods and wire.

The segregation of K in intercrystalline surfaces is lowered with the increase of the total boundary area (lowering the grain size). In this process the dimension of K pores arising at high-temperature annealing is lowered too.

An estimation of changing the density and the relative volume of rods and wires in the course of processing was carried out with the help of differential hydrostatic weighing [1].

For the tungsten rods manufactured by four-roll rolling a new kind of texture(that was not observed earlier) was revealed. In this case there was



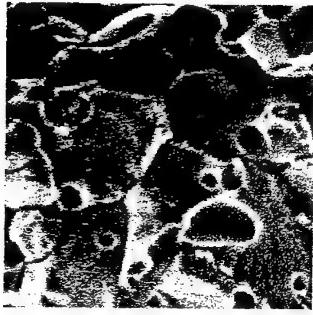


Fig. 1. Fracture surface of a tungsten billet, x2500



Fig. 2. Fracture surface of a tungsten rod 7,5 mm in diameter, x5000

observed the axial texture $<110>$ and limited texture with $\{001\}<110>$, components which is formed by each pair of rolls in their rolling plane. The following annealing of such rods at 2200C leads to grain re- orientation with vanishing the limited texture and the intensification of the axial texture $<110>$.

All kinds of plastic deformation of tungsten billets investigated in the presented work: rotation forging, screw rolling and four-roll rolling form a cellular dislocation substructure.

The singularities of influence of an aspect of the technological scheme of production tungsten rods (rotation forging, screw rolling, four-rolling, as well as the availability of intermediate annealings) on densification rods, fracture toughness K_{Ic} , structure, texture are established

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METHODOLOGY USING ACOUSTIC METHODS FOR THE
DETERMINATION OF ELASTIC CHARACTERISTICS FOR SINTERED
MATERIALS

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At present sintered materials have wide applications [1]. At revealing their operation possibilities one of the most important characteristics is elasticity [2]. To determine elastic characteristics for cast materials acoustic methods are used effectively where elastic modulus are calculated within the limits of an elastic theory for an ideal solid body by elastic waves velocities measured experimentally [3].

Variety of sintered materials architectonics [4-6] and high sensitivity of acoustic fields to the solid body structure [7] complicates the connection between the measured and sought characteristics [8] and leads to the large variety (depending on phases concentrations, their physical properties, morphology, mutual location, degree of coherency and so on) in variants of problems for determining those materials properties using acoustic methods [9]. Therefore transferring traditional ways for determining elastic characteristics of cast materials using acoustic methods to the sintered materials is not always possible, and a question of peculiarities at solving every such a problem needs a special consideration.

Methodology for the use of nondestructive acoustic methods to determine elastic characteristics of sintered materials is proposed. In accordance with this methodology it is necessary to perform the following chain of operations for each sort of material: to conduct analysis of an investigated object choosing the level of a phenomenological approach to the investigated physical phenomenon; to construct a material structure model at the accepted level in detailing the investigated object; to construct the material mechanical model with a mathematical description of an acoustic field in the material and a choice of the diagnostic parameters connected with the investigated phenomena; to ground (to work out) the acoustic method for investigations with the choice of measured parameters; to perform the apparatus and methodical realization of the method providing a required accuracy of measurements; to determine measured parameters experimentally; to determine the material acoustic characteristics by parameters measured; to determine (to estimate) the material properties by acoustic characteristics. It was shown that in the main an achievement of successes at solving the problems for each concrete material was conditioned by possibility of obtaining (on the base of a proper choice of the level for the phenomenological approach at constructing the structural and mechanical or empiric models adequate



to the conditions of this problem) the correlation dependences between the parameters of acoustic and controlled material characteristics in view of physical properties, sizes, form, relative fraction, quality of intercommunication between the structural elements at the chosen phenomenological level. It was marked that the important condition for obtaining the correct results was accounting the peculiarities of acoustic fields in the heterogeneous mediums by the proper choice of methods and techniques for measurings and also measuring equipment.

Problems for determining elastic moduluses in the porous and composite powder materials were analysed on the base of the methodology proposed. It was supposed that the materials structure models were biophase systems in which the elements of each phase had the simplest geometrical form and were distributed within the bulk of the material uniformly, and variables were relative phases concentrations (porosity) and sizes of their elements. Relations between elastic waves characteristics and materials elastic moduluses were proposed within the limits of each model. The calculation relations were compared to the experimental results.

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**APPLICATION OF THE FINITE ELEMENT METHOD AND THE
STRAIN THEORY FOR PLASTIC TASK DEFORMATION OF POROUS
MATERIALS**

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The main mathematical apparatus for elasto – plastic task solution within simple loading is the theory of little elasto – plastic strains. Absence of differential equations is main cause of its wide using.

Equations for porous materials were elaborated under the theory of little elasto-plastic strains. The equation define influences of stress invariants on strain invariants as independent from each other. All material constants and function could be defined for some range of porosity. This allows to make all counting more simple.

Experimental confirmations of elaborated equations had been performed. Given equations had been used in Finite Element Method of deformational type. FEM analysis for plane and axissymmetrical tasks had been done.



EXPERIMENTAL RESEARCH OF A CREEP AND LONG-LIVED STRENGTH SINTERED POWDER MATERIALS

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On the given time there are by unified, affirmed experience criteria of an estimation of processes of a creep and long-lived strength of powder materials. It is connected to a number of difficulties, which one include: technology of obtaining of prototypes of a different porosity with stable mechanical properties and complexity of realisation of experiments.

The activity on an experimental research includes mining a technique of realisation of experiments, technology of obtaining of prototypes of a different porosity, creation of a prototype installation for researches. It will allow performing experiments in a broad band of temperatures at simple and composite states of stress.

Having realised these points, the experiments on research of a creep and long-lived strength sintered of powdered materials from a powder IIMC, porosity 6 - 19 %, ПДКР - 15 - 30 % are performed, at simple and composite states of stress. Influencing a porosity, temperature, type of stress state on a strain rate of a creep and long-lived strength is investigated. The experimental data have allowed to reveal a number of the special criteria of a creep and long-lived strength, namely: nature of the charts of a creep and long-lived strength depending on a type of stress state, destruction of stuffs and many other, which were not investigated before.

The further experimental researches will allow to test the hypothesis about existence of a potential of strain rates of a creep, which one probably coincide by the form a potential of speeds of plastic deformation, and also hypothesis about existence of a surface of a creep.



OTHER PROSPECTIVE MATERIALS

213-356



THE ORIGIN OF AN ANOMALOUS TEMPERATURE DEPENDENCE OF THE YIELD STRESS OF AN ORDERED CuAu ALLOY

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An irreversible positive temperature dependence of the yield stress $\sigma_{0.2}$ was detected for an ordered CuAu alloy, which had a single-domain grain structure in the initial state. To obtain that structure, a disordered alloy sample, which was deformed beforehand by rolling to 72-90%, was annealed at T (350°C) $< T_c$ to realize a complex discontinuous reaction of ordering and then was cooled to room temperature at a rate of 10 degrees per day. The alloy structure was fully recrystallized. Each grain represented a single domain. Grains did not contain a lamellar structure. The internal structure of grains was extremely homogeneous: no variation of the diffraction contrast was observed in electron microscopic images. However, when the alloy was heated again to 350°C and cooled quickly to room temperature at a rate of 100 degrees per minute, grains had a fine structure including nanosized regions 6 to 8 nm in diameter with different directions of the c -axis. Those regions alternated in the grain bulk. Grains ceased to be single domains.

When the initial foil was heated in an electron microscope, local disordered regions were formed in the structure of a single domain. "Scintillation" of the structure began at 250°C and was enhanced up to 300°C . That effect probably reflected a dynamic nonequilibrium of the structure: the initial structure of single domains was restored when the sample was cooled to room temperature. Heating to a higher temperature led to formation of stable nanosized regions of a disordered phase in the grain bulk and growth of their volume fraction. The size of the disordered nanoregions, which appeared upon heating above 300°C , exceeded the critical value. The detected changes in the CuAu structure, which were analogous to the aging process, caused the increase in the yield stress of the alloy with elevating temperature, i.e. led to a positive temperature dependence $\sigma_y(T)$. The yield stress was a maximum at 280 to 300°C . When the temperature was elevated to 315°C , the yield stress dropped and plasticity increased insignificantly. Higher temperatures caused embrittlement and premature failure of the alloy, while σ_y was not reached.

A thermoresistometric analysis confirmed that positive temperature dependences of the yield stress in Cu_3Au and CuAu had the same origin.

The origin of the anomalous temperature dependence was inadequate to temperature strengthening of these intermetallics.



ON THE ROLE OXIDATED OF THE DEFORMATION AND TOUGHNESS OF ELASTOMERS

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The kinetic conception of the polymer mechanical oxidation strength is considered as thermofluctuation process of the strained chemical bonds breaking with formation of the end free radicals. It is supposed that their interaction with oxygen results in formation of the end C=O groups. The physical model of the primary centres (hotbeds) of breaking in the mechanical strained (stress-strained) samples was developed according to kinetics of the accumulation of the such C=O groups.

However, our experiments have shown that the phenomenon of polymer oxidation under the action of strain is the complex process, which depends both on the sample purity degree and on the polymer structure before its fracture state. It was found that oxygen containing groups completely is "washed" at the washing of the oxidized samples in the organic solvent. These facts were explained as follows:

1. The oxidation of mechanoactivated (stress-induced) radicals is accompanied by decomposition of macromolecules on the small pieces or fragments, which have been named "microsplinters".

2. It is established that C=O groups have included into the chemical composition of the microsplinters, but are not bound at the ends of the breaking macromolecules.

Therefore, the supposition that at the oxidation of polymers the established C=O groups are the end groups is not correct. It is showed that the intramolecular oxidation reactions lead to formation of microsplinters with aldehyded type of C=O groups. In the samples where the reaction occurs by way of exchange of interchain free valencies, the ketone type of C=O groups is formed principally. It is established that the concentration of microsplinters, containing C=O groups, is increased with the decrease of oxygen pressure in the range 20-730 mm Hg.

The formation of new primary organizations of supermolecular structure with the oxidation simultaneously in the stress-strain polyisoprene have been found.



RECOVERY AND RECRYSTALLISATION IN MICROCRYSTALLINE METALS AND ALLOYS PREPARED BY INTENSIVE PLASTIC DEFORMATION METHODS

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At present work the results of experimental and theoretical researches of structure stability and mechanical properties for nano - and microcrystalline (NMC) FCC metals and alloys obtained by equal-channell angular pressing (ECAP) are performed.

It is experimentally established, about recrystallization temperature in NMC metals the grain growth processes have anomalous character - only separate grains are essentially drew, and during annealing volume fraction of these anomalous grains is increased under exponential law.

Analisis of mechanical properties in NMC metals and alloys at low-temperature annealing shows the beginning of recrystallization process is accompanied by the process of some hardening. And besides the hardening is more visible at the temperature-time dependences on a macroelasticity point than at the temperature-time dependences on a yield point and microhardness.

It was investigated the influence of the small impurity addition (up to 0.1%) on the value of recrystallization temperature and the character of grain growth.

The models of defect structure evolution processes in NMC metals and alloys at recovery and recrystallisation conditions are offered. The models are based on the theory of the non-equilibrium grain boundaries.

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THEORY OF NON-EQUILIBRIUM GRAIN BOUNDARY DIFFUSION AND ITS APPLICATION TO STRUCTURAL SUPERPLASTICITY

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There was carried out theoretical investigation of the structure, thermodinamical parameters and diffusion characteristics of equilibrium and non-equilibrium high angular grain boundaries (GB). It is explained that the boundary structures may be described using the «islands model», and that the free volume is the main parameter characterizing distribution of the islands and, correspondingly, the boundary structural state. There is proposed a new model of GB diffusion. The model is based on the ideas of heterophase fluctuations when there are changed the size of the islands having high diffusion. There were carried out the calculations of enthalpy, entropy and GB diffusion activation energy for some metals and alloys. There is described the structure and the properties of non-equilibrium GB. There is explained that anomalies in diffusion parameters and thermodynamical characteristics of non-equilibrium GBs are the result of their free volume increase. The reason of this change is the expense of the free volume introduced by lattice dislocations get into the boundaries and spread there.

In the work there is analysed the role of the strain-enhanced diffusion in the GB processes of superplasticity. There are discussed the peculiarities of processes of the grain boundary sliding, strain enhanced GB migration and the grain growth.

There is in detail considered the GB free volume value influence on basic superplastic flow parameters and microstructure evolution for different types of superplastic materials having small and large GB free volume.

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**EXPERIMENTAL AND THEORETICAL INVESTIGATIONS OF
INTERNAL FRICTION IN MICROCRYSTALLINE METALS PRODUCED
BY EQUICHANNEL ANGULAR PRESSING TECHNOLOGY**

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The investigations of internal friction in microcrystalline copper (99.98 wt.%), nickel (99.91 wt.%) and chrome bronze (Cu - 0.38 % Cr) were carried out. The materials studied had special microstructure with characteristic size of the structural elements $\sim 150 \div 200$ nm. This structure was obtained by equichannel angular pressing (ECAP) technology. Internal friction were measured by the resonance acoustic technique at the frequency ~ 1 kHz and inverted torsion pendulum at the frequency ~ 1 Hz. The internal friction measurements were carried out during heating and cooling at a rate of 2 $^{\circ}$ C/min within the temperature range of $20 \div 500$ $^{\circ}$ C and isothermal annealing at the temperature range from 100 to 300 $^{\circ}$ C. The investigations have shown that the internal friction Q^{-1} in microcrystalline materials essentially differs from the internal friction values of coarse-grained metals at the same range of temperature. The experimental study has shown an existence of the peaks and other anomalies on the temperature dependence of internal friction as compared to the pure coarse-grained metals. During isothermal annealing of copper and nickel new internal friction peak was found.

We assume that the nonmonotonous behaviour of internal friction curves can be explained by the dislocation and grain boundary models of internal friction modified for specificity of microcrystalline materials (the small size of grains, the high density of lattice dislocations and the nonequilibrium state of grain boundaries). At the frequency ~ 1 kHz internal friction is controlled by the thermally activated motion of dislocation kinks. At the frequency ~ 1 Hz internal friction in microcrystalline metals is caused by joint action of two mechanisms: dislocation mechanism and grain boundary mechanism of internal friction.

The proposed models of internal friction are rather effective for describing of internal friction in microcrystalline materials produced by ECAP technology.

The authors thank for support International Scientific-Technical Center (ISTC) (Grant № 1413) and also the program "Basic Research in Higher Education" (BRHE) and Research and Educational Center of Scanning Probe Microscopy (REC SPM) of Nizhny Novgorod State University. This research was supported in part by Grant № 00-02-16546 from the Russian Foundation of Fundamental Research (RFFI).



THE MECHANISMS OF THE DEFORMATION LOCATION IN IRRADIATED MATERIALS

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The brief review of the physical models of plastic deformation localization phenomena in the temperature range including the low temperature radiation embrittlement effect ($T \leq 0,35 T_m$) has been done.

These phenomena classification is proposed:

1. The stationary dissipative structures (dislocation channeling);
2. The moving fronts of deformation localization (the Chernov-Luders bands type);
3. The macroscopic bands (the Danilov-Zuev relaxation waves type).

The new approach for the dislocation channeling mechanism description on the base of considering the collective behavior of dislocations and their interaction with radiation defects is proposed [1]. It is shown that the fraction on dislocations overcoming radiation defects with high velocities in the dynamical regime increase with increasing of radiation hardening. When dose increases the dynamic regime of deformation occurs at lower velocities compared with the unirradiated materials.

The model of collective behavior of dislocations in irradiated materials is proposed on the base of the kinetic equation for the dislocation density with taking account of Burgers type nonlinearity [2]. It is shown, that increasing of the long-range internal stresses due to the markable increase of dislocation density in the Chernov-Luders types of bands. As results, under doses ≤ 1 dpa, dislocation density in the bands increase to ten times.

On the base of the kinetic equation of dislocation density evolution in the irradiated deformed material the possibility of the localized structures development of relaxation waves Danilov-Zuev type has been shown [3]. It is also shown that the localized structures form faster with the irradiation dose increase and these structures became more relief under such conditions. The possibility of localization effects (embrittlement) decrising by the plastic deformation micro-level activation is demonstrated in relation with the thermal activated processes.

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TO A TECHNIQUE OF DEFINITION OF THE MECHANICAL CHARACTERISTICS STRUCTURALLY HETEROGENEOUS MATERIALS AT COMPLEX LOADING

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Poor quantity of experimental data in stress space, in other as well as in space of deformations, is connected to lack of universal testing machines on complex loading. And though in engineering practice processes more composite, than processes of mean curvature, the spline processes or their combination meet, apparently rather seldom. Research and experimental construction of functionals of a plasticity for such paths is of interest from the idealized and practical points of view.

At realization of experiments is by an urgent question the definition of the power and deformation characteristics during loading with some step. As the process of loading is model there is for short period that definition of these characteristics by visual methods it is impossible.

For research of behavior was model at the miscellaneous schemes of loading by us the universal exploratory complex consisting from the installation high-pressure is designed, which one allows to conduct experiments in pathways P-q, P-M, P-q-M and measuring complex on the basis tensor measuring station СИИТ-2. Tensor measuring station allows to plug about 500 probes and to view on an information diagram board.

For process control of measurement and preservation of the data received from tensostation the software - hardware complex for the coordination tensostation from a computer was designed. The program TENZ allows to select indispensable quantity of sensors, to test them, to set quantity and speed the reference to each probe, after realization of experiment to deduce outcomes on a screen by the way of charts in non-dimensional values or to keep on a computer. The program TENZPER under the calibration charts translates observed data in real values of stress and strains.



DEVELOPMENT OF NEW HIGH-STRENGTH CUSTING ALUMINUM ALLOYS WITH THE USE OF THE COMPLEX STRENGTHENING MECHANISM

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A new high-strength eutectic alloy in the ternary Al-Ge-Mg system was created. This alloy is considered as an alternative to silumin alloys for low temperature working conditions ($20-150^{\circ}\text{C}$). The alloy has high strength and satisfactory plasticity which are realized due to the complex hardening mechanism.

The basic alloy composition is located nearly the eutectic on the quasi-binary section of the ternary Al-Ge-Mg phase diagram. The composite strengthening mechanism is provided by the fiber-lamellar structure of $\alpha\text{-Al}+\text{Mg}_2\text{Ge}$ eutectic (pic.). Precipitation and solid solution hardening are realized by means of the rational alloying and optimal heat treatment. Precipitation hardening is activated by Zn-containing phases, which are precipitated under heat treatment. Optimal heat treatment conditions give the possibility of dissolving primary phases without coalescence and coarsening eutectic structure.

Using of the complex strengthening mechanism as mentioned above allowed to obtain

the record strength of developed AGM-alloy from the comparison with an industrial casting aluminum alloys (tabl.).

Tabl. Mechanical properties of high-strength cast aluminum alloys

Alloy, country (reference)	Tensile properties range at 20°C		
	$\sigma_{0.2}$ (MPa)	σ_B (MPa)	δ (%)
Alloy AGM, Ukraine	490-620	540-660	1-2
Cast alloys, USA (Aluminum and Aluminum Alloys, 1993)	200-414	278-476	2-6
Cast alloys Al, Russia (MiTOM, 1993, N7)	200-500	240-550	1-12



Pic. Microstructure of the eutectic alloy of the Al-Ge-Mg ternary system.



ON THE RESISTANCE of LOW-ALLOY CHROMIUM ALLOYS TO DEFORMATION WARM WITH ROLLING

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The important technical characteristic of rolling is resistance to deformation, which knowledge allows to establish stress, that is necessary for plastic deformation.

The researche show, that the resistance to deformation is essentially reduced with increasing of temperature from 473 to 773 K. When temperature grows from 473 to 773 K resistance to deformation irrespective on deformation at least for (lnl_0/l_i) 0,05; 0,4; 0,19 from 440 to 380 MPa independently reduced approximately for 100 MPa.

Increasing a degree of deformation from 0 up to 0,23 results in continuous growth of resistance to deformation, which reaches at temperature of heating 573 K 440 MPa, and at 773 K 380 MPa independently on temperature of tests. More intensive gain is observed in the beginning of deformation. It takes 90-120 MPa deformation of 0,05, and at deformation in an interval 0,18-0,23 (also on size 0,05) the gain of resistance takes only 15-20 MPa, but in the last case the higest resistance to deformation (400-500 MPa) may be found. Increasing rolling speed does not result in essential growth of resistance to deformation. E.g., increasing speed for two orders (from 0,1 up to $10 s^{-1}$) the resistance to deformation grows from 320 to 365 MPa and this dependence is close to linear.

Investigation of technological plasticity by a method of wedge rolling temperature of rollers 373 between and 473 K has shown that the limit of plasticity is plastisity equaled to zero in an interval of temperatures of heating 293-373 K bat it is 61-62 % at temperatures 573-773 K. Increasing temperature of rollers for 100 K allows reaching doformation degree per one passage for 66-83% at the working temperature named. Increasing working temperature to 973 K leads to decreasing degree of deformation preseadingfracture due to deformation aging taking please in low alloyed Cr-0,5La-0,2Ta alloy at this condition. In low-alloy chromium (Cr-0,5La-0,2Ta) the increase of temperature of rolling up to 973 K is attend to decrease of a degree of deformation before failure because of deformation aging.

Thus, this study confirms an expediency of rolling chromium blanks at 773 K with heating of rollers up to 523-573 K.



APPLICATION OF NEW METHODS OF THE CONTROL AT VIBRATIONAL SEAL OF POWDERED MATERIALS

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The wide development of a powder metallurgy in a direction of manufacturing of high-precision parts provides useful increasing of production volumes of such items in engineering. For manufacturing of metallic and nonmetallic items from bulk materials often will use a method of vibration seal. This way allows using the simple equipment for obtaining items of the miscellaneous shapes with high uniformity of volumetric distribution of density. The vibration formation of powdered materials has following advantages: the costs of a press - tool is descended; elastic aftereffect is reduced to minimum; the achievement of high and even density of the obtained items is provided, and also the effort of pressing decreases.

The principle design concepts of miscellaneous types of the vibration installations are well known and are described in the literature. But in some cases recommendations are not exact. The main problem of its elaborations is absence of common method of main process parameters registration and adjustment. The practical uses are introduced by creation of the universal vibration machine, which would not demand considerable changes in a design of a normal equipment, besides, the installation should provide full regulation and control of parameters of chattering, to have enough broad band of regulation, small energy output. In LSTU the experimentally - exploratory machine was elaborated, which responds given conditions. The control and adjustment of parameters at vibration seal in the offered installation implements the automated measuring complex, which based on computer. Complex allows to fix main specifications of process of vibration pressing, to use sensors of different designs without signal conditioning units, to receive average values of parameters for given period, to select optimal modes of pressing for different stuffs.



INTRODUCTION OF HELIUM IN TIN WITH DISLOCATION-DYNAMIC DIFFUSION METHOD

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The present work pursues investigations [1,2] of the physical nature of the phenomenon of dislocation-dynamic diffusion {DDD}. The essence of the DDD phenomenon is penetration of particles of gaseous or liquid medium into crystalline bodies along nucleated and moving dislocations in the course of plastic deformation. The content of helium introduced in tin monocrystals {99,999%} of [110] orientation by tensile deformation to different strains at T=4,2 K by mass-spectra method was measured. The measurements revealed the presence of helium in deformed tin and the increase of helium content with the amount of deformation. The strain dependence of amount of helium extracted correlates qualitatively with the stress-strain curves. Spectra of helium extraction from deformed specimens during their dynamic annealing at T=300-800 K were obtained. Narrow peaks of helium extraction for temperatures near the melting point of tin were found. They indicate the presence of deep traps of helium presumably chemically related with defect structure of metal. Mechanisms of penetration and extraction of helium from tin deformed in liquid helium medium are discussed.

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OVERDAMPED PHONONS AND SYNERGETICS OF ELECTROPLASTIC DEFORMATION: FATIGUE

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In this paper the small-cycle fatigue of steel is examined. It was shown by experiments [1] (by application of ultrasound diagnostics technique) that one can prevent the fatigue damage by the treatment of details with the help of electric impulses of large amplitude. This treatment is accompanied by increasing the resource of durability in 15-25%. The typical parametres realized in testing are indicated in [1] (the fatigue limit, registered by abrupt fall of ultrasound velocity [1], frequency of current impulses the time of current action on specimen, amplitude of current value, critical external load, frequency of mechanical impulses). The description of fatigue in terms of plastic deformation (Manson's law [1]), but not the law of stresses is used by us. The effect of increasing the resource of durability mentioned above is linked by us with the reconstruction of dislocation mode of plasticity: the more active – the synergetic one – by the role of mobile dislocations in contrast with the immobile ones [1]. The conception of overdamped phonons (with imaginary law of dispersion) [2], showing the indicated reconstruction of dislocations ensemble and the conception of electronic wind [1] enable to carry out the quantitative analysis of experimental data. These data are not badly described by equation for effective stresses of electronic wind $\sigma = (mV_F/e)j$; [1], where m is a mass of electron, V_F is a velocity of electron on Fermi surface, e is the charge of electron, j is the density of current.

It is necessary in connection to discuss first the synergetic role of mobile dislocations [3]. Not going in details of rheological scheme for the displacements and plastic deformation developed in this work, it is important to note that the putting equal of effective stresses of electronic wind to the stress of arising of elastic-plastic region in ideal crystal (for example, dislocation) [2] quite correctly shows the synergetic role of mobile dislocations. From the point of the conception of overdamped phonons the synergetic role of mobile dislocations is expressed in a specific nature of the process of the arising and dynamics [2,3].

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DISLOCATION MODE OF ELECTROSTIMULATED FATIGUE AND OVEREDAMPED PHONONS

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In this paper the small-cycle fatigue of steel and the dislocation formation is examined. It was shown experimentally [1], that by means of details treatment with the help of impulses of large amplitude one can prevent the fatigue damage (the ultrasound diagnostics of materials was used). Such a treatment is accompanied by increasing the resource of durability in 15-25%. The typical parameters being realized in testing are enumerated in [1] (fatigue limit registered by sharp fall of ultrasound velocity [1], frequency of current impulses, time of current action on specimen amplitude value of current critical (external load, frequency of mechanical impulses). The description of fatigue in terms of plastic deformation (Manson's law), but no the law of stresses is used by us. The effect of increasing the durability resource examined by us is linked with the reconstruction of dislocation mode of plasticity: by more active role of mobile dislocations as opposed with the immobile ones [1]. The conception of overedumped phonons (with imaginary law of dispersion), showing the indicated reconstruction of dislocations ensemble and the conception of electronic wind [1] allows to carry out the quantitative analysis of experimental data. These data are described by equation for effective stresses of electronic wind $\sigma = (mV_F/e)j$; [1], where m is a mass of electron, V_F is a velocity of electron on Fermi surface, e is the charge of electron, j is the density of current. As for the details of reconstruction of the dislocations ensemble, it can be investigated with the help of comparison of effective stresses of electronic wind with stresses of the arising of elastic-plastic region in ideal crystal [2] (the detailed discussion for dislocation is given in [3]). The conception of overedamped phonons has the direct relations to the similar type of theoretical notion about the structural reconstructions in materials with not ideal crystalline structure [2]. Unlike the materials with almost fully overedamped phonons (synergetic region with imaginary spector of phonons in almost the whole region of wave vectors [2], that occurs in high densities of dislocations [4], the region of effects being of interest from the point of electroplastic deformation corresponds to weakly overedamped phonons [2,4].

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METAL PLASTICITY DURING STRUCTURE FORMATION BY TWIST EXTRUSION

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Severe plastic deformations at last time attract intent attention of researchers that work in field of physical materials science. It's a powerful tool for transformation of materials structures. Basic structure forming regularities during the plastic deformation are determined by the parameters of initial structural material state, temperature-velocity deformation conditions, and mechanics of deformation process. The last is specified by scheme of external kinematics and force action on deformed volume (by the deformation scheme, tools geometry, conditions of contact friction etc.) and determine the series of basic process parameters: value of accumulated deformations intensity and deformations velocities in any spots, orientation of macroshear spatial evolution, configuration of deformed volume changes, homogeneity of stress and strain distribution etc. Each of these parameters influence in the corresponding way on formation of structure and properties of work materials.

There are two ways are used now for deriving severe plastic deformations without forming. The first way is torsion of disc-shaped blanks in Bridgmen anvils and the second one is equal channels corner extrusion of cylindrical and prismatic blanks.

A new method for deriving severe plastic deformations by hydromechanical extrusion of prismatic specimens throw a twist channel die is offered in this paper. All the cross-sections of the channel that is orthogonal to the extrusion axis are invariable along this axis. An inclination of twist line changes along the matrix height and on initials and final section it's equal to zero. The peculiarities of channel geometry bring that the form of specimen doesn't change. It allows multiple extrusion of one for the purpose severe deformations accumulation.

A change of dislocation arrangement and physicomechanical properties attached to twist extrusion is explored. The energy-power parameters and a value of accumulated deformation after one skip are calculated. The upper estimate for the toughness resource spending quantity is carried out. That allows estimate a maximum value of the skips without specimen failure. The suggestion on industrial use of the scheme for deriving unique physics and mechanical characteristics of bars from metals and alloys is advanced.



FRACTAL MODEL OF ANDRADE CREEPAGE

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The mathematical model of non-homogeneous creepage which is based on the fractal imagery of non-equilibrium processes has been developed[1-2]. The analogue of Taylor's series and differential-integral operators for fractal functions have been calculated. The connection between fractal dimension of time relaxation and parametrers of Andrade's creepage has been defined.

The solution for differential equation with operators of fractional order for system deformation $\varepsilon(t)$ has been given

$$\varepsilon(t) = [s_0 + s_\infty \cdot \sum_{n=0}^{\infty} \frac{(-1)^n \cdot \left(\frac{t}{\tau}\right)^{\alpha(n+1)}}{\Gamma[\alpha(n+1)+1]}] \cdot \sigma, \quad (1)$$

where $\Gamma(x)$ is the gamma function; s_0 and s_∞ are the matters features, t -time, τ - time of relaxation.

If you are limited by the first members of the series (1), then the dependance will be the following

$$\varepsilon(t) = \varepsilon_0 + a_1 \cdot t^\alpha + \dots, \quad (2)$$

where the first member in (2) ε_0 defines the creepage not depending on the time and the second member does the same, for the creepage.

The result given is in a good agreement with Andrade's dependance. The fractal series of relaxtion time is shown to have dimension $1/4 < d < 2/3$.

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ELASTIC PROPERTIES OF COMPOSITES WITH CHAOTIC STRUCTURE

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In this communication, some quantitative results of modelling of elastic properties of composites with chaotic structure are presented. The developed by the authors numerical iterative method of averaging is used, based on results of fractal geometry and on renormalization group approach [1-4].

The method exploits averaging within a "blob" model in which the tensor character of elastic properties is preserved [5,6].

The presented method of building structural models for chaotic systems and the iterative method of determining their elastic properties show a good agreement with numerical methods of modelling the elastic properties of percolating systems. The method can be used not only for predicting elastic properties of model percolating systems (e.g. with infinite elastic constants) but can also be used for describing elastic properties of real materials composed of phases of finite elastic properties.

The critical indices of the bulk modulus, K , in the elastic ($\tau=3.200\pm0.002$) and super-elastic ($s=0.62962 \pm 0.00002$) region have been calculated. It was also shown that when the concentration tends to the critical one, p_c , then the ratio K/μ tends to $4/3$, where μ denotes the shear modulus. This result implies that the Poisson's ratio does not depend on properties of the original phases at the percolation threshold and is equal to $1/5$.

Comparison of calculated numerical data gives a good agreement with experimental results obtained for porous metallic materials.

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ACOUSTICS EMISSION DIAGNOSTICS OF OBJECTS

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Diagnostics of working constructions assumes an evaluation of their carrier capability. As a rule, in a construction the safety factor is mortgaged, this method is only based on the mechanical approach to an evaluation of serviceability of technical objects.

The latest reaching in the field of physics of the strength of solids led to studying termofluctuation nature of destruction of the loaded solid. It in turn has reduced in understanding, that the macroscopic destruction not only is connected with reaching of strength, and is a final stage in long process of accumulation of microdamages. The first stage of this process is connected to accumulation of microscopic defects or micro cracks, second – in emerging of the dangerous center of destruction after reaching of threshold concentration of micro cracks. As a rule, the first stage takes much more time, rather than second. Inspecting process of accumulation of damages by a method AE it is possible to define what resource (safe life) it of serviceability and on what stage there is a researched object.

Is valid, each microtupture or the dynamic structural reorganization results in a radiation of elastic impulse. Amplitude and pulse duration are connected to a size of forming cracks, and the modern systems (devices) allow to register formation of microcracks $\sim 10\text{-}20$ microns, that much exceeds a resolving power of ultrasonic flaw detectors.

Besides the method allows to perform monitoring all objects in whole, locating a place of formation of microcracks and appearing dangerous centers of destruction. The important reaching of an offered method is the operating analysis of several parameters AE. This analysis is bases on fundamental scientific researches of process of a radiation of acoustic impulses when formatting microcracks, main regularities of process of accumulation of damages, and also features of shaping and forming site of origin for fracture. It is difference our method from being available.

The kinetic theory of fracture, mathematical statistics when analyzing acoustics emission streams, computer modeling of processes of accumulation and the integration of cracks have allowed significantly to advance this method.



MICROSCOPICS of THERMOELASTIC EFFECT in SOLIDS

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Thermoelastic effect results in the change of materials temperature while their elastic adiabatic loading. Thermodynamic description of the effect is given by Kelvin's formula according to which the variation of temperature is proportional to mechanical stress and thermal expansion coefficient.

There are two peculiarities of thermoelastic effect:

- temperature variation may be of the different sign for materials with different structure
- if this sign is positive then the increase in heat energy of material exceeds the work of mechanical loading.

For materials with different structure different thermoelastic mechanisms are examined.

For small molecular materials (metals, ordinary crystals) the anharmonic oscillator is chosen to be the main element of their intrinsic dynamics. It was shown theoretically and proved numerically that mechanical loading of such an oscillator causes variation in its kinetic energy, while the potential energy variation occurs with opposite sign. A simple model, explaining main features of thermoelasticity for these types of materials is proposed.

For macromolecular (polymer) materials conformational dynamics is discussed, giving rise to special features of thermal expansion and energetics of loading in these materials. Good agreement of calculated and experimental results is achieved.

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STUDY OF SELF-ORGANIZATION AND DYNAMIC CHAOS OF THE
MEZOSCOPIC STRUCTURE RELAXATION PROCESSES IN PLASTIC
DEFORMATION NONMETALLIC SOLIDS BY A NEW
ELECTROMAGNETIC METHOD

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Electromagnetic wave radiation during mechanical deformation some nonmetallic solids and minerals ($A_x B_{8-x}$ single crystals, ice, ceramics, marble and etc.) was investigated. It was the sequence of large amount (10^3 - 10^4) pulses of electric field potential near the deformed sample, called electromagnetic emission (EME). We use a alkali halide crystal and ice as the model substances. There are three groups EME pulses: I-pulses have front duration $t_f < 100$ ms it was caused by dislocation pile ups (consisting to $\sim 10^2$ dislocations) dynamics; II-pulses with $t_f \sim 1$ - 10 μ s caused by growth of cracks ≥ 10 μ m length and III-pulses with the $t_f < 100$ ns caused by dielectric breakdowns between different points of solid surface. It was shown that the sum of I-pulse amplitudes is proportionate to degree of plastic deformation and the sum II-pulses amplitudes is proportionate to the area of the fracture surface. Thus, statistical analyze of ensembles of I- and II-pulses allow to get the information about plastic deformation and fracture kinetics on the mesoscopic structural level.

Autocorrelative characteristics of EME is determined by coefficient of correlation between the pulse amplitude and the following pause duration that is a quantitative measure of correlation between of structure relaxation events that caused this pulses. It was found the interchange of self-organization and chaotization of events of the plastic flow process of the dislocation slip band level. We suppose that the some stage of structure relaxation is realized by self-organization of corresponding process on the previous stage.

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**FORMING ACCUMULATIONS OF DISLOCATIVE LOOPS AND THEM
PART IN A DESTRUCTIVE OF BRITTLE MATERIALS**

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Experiments showed that the destructive process under ultrasonic action (UA) for brittle materials had different from one for metals. The destruction of brittle materials is determined forming accumulations of dislocative loops which have density about $10^{12} - 10^{13} \text{ cm}^{-3}$.

The analysis of vacancy processes showed forming regions of vacancy oversaturation near free surface and grain boundaris under UA. These regions are formed diffusion of vacancies from surface and grain boundaries which are powerful vacancic sources.

Dislocative loops are appeared from surplus vacancies in regions of vacancic oversaturation. So surplus vacancies are disappeared. Thus it is to some degree "vacancy pump". The density of dislocative loops is increasing in these regions under UA and so is the value of internal stress,which is made these loops, is increasing too. For brittle materials the calculation show when density of loops is about $3 \cdot 10^{12} \text{ cm}^{-3}$ the internal stress more a destructive value and materials are destroied.



INFLUENCE OF HYDROSTATIC PRESSURE ON PHYSICAL-MECHANICAL PROPERTIES OF A MOLTEN ALLOY Al-Si6-Cu4

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The aluminium alloys are widely used to manufacture of details of automobile engines (pistons, head of cylinders etc.). As well as all molten alloys, they contain a fair quantity of foundry micro- and macropores that essentially reduce density, aggravate elastic and plastic properties, that in a complex reduces the operation safe life of alloys. There are a lot of papers dedicated the nature and methods of elimination of a crystallising porosity in metals and alloys [1,2]. However physical mechanisms of foundry pores healing in hydrostatic squeezing (HS) conditions are not quite clarify, and also there are no guidelines at the choice of optimum regimes of treatment, which one would give in essential diminution of a porosity and its inhomogeneity.

The aim of the present work was the examination of influence of HS at different pressures (p) and temperatures (T) on density (ρ), velocity (V) and signal attenuation (α) of ultrasound (US) and mechanical performances of polycrystalline silumin Al-Si6-Cu4. HS was carried out in thermohydroextrusion installation TGE-20P, permitting to vary a level of p from 0 up to 800 MPa in temperature range T = 20-700°C. For a hydraulic medium was an 50% intermixture of glycerine with water. The precision of the check of pressure conditions was ± 10 MPa, temperature measurements accuracy was $\pm 10^{\circ}\text{C}$. The installation for measuring of density (ρ) and US studies was the same as in papers [3,4]. The density (ρ) was measured by method of a volume determination based on Archimedes' principle with usage of the quartz or germanium ethalons with the instrumental error of density measuring 0,01% for specimens with mass 10g. The instrumental error of measurements of absolute values of velocity V on the basis of 10 m/c was 0,01% and rational ones – 0,001%. The measuring error of α at sample length 10 mm was 1-3%. The US measurements were conducted by longitudinal waves on frequency 10 MHz and width of ultrasonic bundle - 2-3 mm and 7 mm. The mechanical testing were carried out by installation Testatron 1288/1004N developed by Wolpert firm.

At US investigation of ingots and initial specimens of an alloy Al-Si6-Cu4 the considerable spatial elastic inhomogeneity and essential difference ultrasonic signal attenuation values of studied objects was detected. The maximum divergence of data on different sites of ingots and studs on V was 10%, i.e. on elastic module $\approx 20\%$ and on α reached 40-45 db/cm, that witness both to a great number of foundry pores, and about their nonuniform distribution in a material. It



was shown that the relative magnification of density ($\Delta\rho/\rho_i$), velocity ($\Delta V/V_i$) and decreasing of signal attenuation ($\Delta\alpha/\alpha_i$) of US after HS the is more, than below middle values of ρ_i , V_i , than above α_i and also than above p and T in the above mentioned intervals of these parameters changes. The correlation under HS between $\Delta\rho/\rho_i$, $\Delta V/V_i$ and $\Delta\alpha/\alpha_i$ was also determined. Lowering of temperature and pressure leads to formation of intermediate states, which characterises by a high level of ultrasonic signal attenuation and acoustic discontinuity of alloy, that is reasoned by incompleteness of compacting (healing) processes. The divergence of data for specimens treated by HS is essentially decrease. It is of particular interest the influence of HS on the mechanical performances of an alloy. Demonstrative ones are the treatments of specimens in regimes with stationary pressure values 400 MPa at different temperatures: 200, 300, 350°C. It is appeared that at low (200°C) and medial (300°C) temperatures a positive increase of a yield stress and strength limit (at 200°C $\Delta\sigma_{0.2} = 16$ MPa, $\Delta\sigma_s = 11$ MPa; at 300°C $\Delta\sigma_{0.2}=22$ MPa, $\Delta\sigma_s = 21$ MPa) occur, while after HS at 350°C the decreasing of these performances in comparison with initial state occurs. The magnification of yield stress and strength limit of alloys after HS can be connected with increasing of dislocation density and diminution of a porosity, that agrees with model of dislocation healing of pores in HS conditions [5,6]. According to this model the diminution of a pore volume that is taking place under squeezing in HS condition happens due to an emission of dislocation loops, that in turn leads after pores healing process finishing to increasing of dislocation density on 1-2 orders. While the temperature of treatment increases the emission of dislocation loops is accompanied by their diffusion dissolution, resulted to increasing of concentration of abundant vacancies, that promotes the polygonization and softening processes of an alloy. In conclusion it should be note that obtained experimental results witness about perspective of employing of used US method for nondestructive testing of industry alloys similarly to Al-Si6-Cu4.

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DESTRUCTION OF POLYETHYLENE PIPES WITH IN-CITE TESTS IN THE CONDITIONS OF NATURALLY LOW TEMPERATURES

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Presented are the results of testing of polyethylene pipes (PE80) with inside pressure until complete destruction at temperatures from 20°C to -44°C. Time, deformation and stress components, temperature, mode and destruction surface were controlled at quasi-static loading. In the said temperature range growth of the limit of forced elasticity and rigidity occurred according to the linear law and perform 2.5-fold growth compared to that at normal temperatures. At normal temperatures the failure area is localized at the top of characteristic swelling. The open system and that of pre-critical cracks are directed perpendicular to the generating one. Factographic investigations show that in the temperature range from 20°C to -18°C the front of open cracks looks like an open funnel characteristic of ductile failure (swelled pipe area is ductile at the top). The same mode was observed at temperatures up to -24°C but without swelling.

At temperatures below -24°C we did not observe the realization of ductile-plastic processes in the material, unlike the material revealed a kind of ductile-elastic deformation. Failure torque of a single crack opening corresponds to the maximum point at the curve $\sigma \sim \epsilon$ and is directed along the cylinder generating. The front thickness of the open crack looks like the funnel cross-section. The latter occurs at quasi-ductile failure: the crack starts at the outer surface and moves towards the inner part of the pipe due to the presence of a defect on the pipe surface. Such dependence of macroscopic failure mode on temperature supports the well known tendency of low temperature change of failure mechanism of thermo-plastic bodies with surface defects.

The most unordinary but the most interesting result was obtained at -44°C. The crack front had the form of the widened cross-section of "inverse funnel", which must have been characteristic of the mode of ductile-elastic cracking. Experiments showed that the destruction occurs by means of formation of a single crack – ductile without swelling and plastic deformations. It is evident that the crack generation occurs in the inner part of the pipe. The said discrepancy suggests that the failure process at this temperature is very complicated and contains at least three different stages. Ductile-elastic mode is dominating. It is possible that the start of plastic destruction occurs at the inner part which is more stressed. More likely it is due to the process of the material recrystallization at local heating. Later the process slows down due to increased rigidity of the outer "layers" and here one can observe the formation of an "echelon" of sequential cracks, whose cross-



section has the form of triangle asteroids, whose upper angle turns to the cylinder normal. As a result of inversion of maximum stresses the crack opening and propagation occurs "by layers". If the pipe pressure grows (the condition of reaching the criterial maximum of hoop stress) one can observe the opening of outer "angles" of the "asteroids". At the stages of sequential destruction we observed the following process: unweakened "walls" between the "asteroids" become longer. The destroyed sample performed them in the shape of characteristic "tabs" at the crack surface. As a result, the failure mode has mixed character, which is confirmed by a complicated shape of the failure surface. Investigation of linear crack showed that beside the elongated parts ("tabs"), the presence of additional surfaces of open cracks, whose direction is parallel to the cylinder generating. They are perpendicular to the basic crack and are symmetrical at the both sides of it (traces of side "astroid" angles).

Locally elongated parts of the material ("tabs") at the first side can be identified as characteristic "elongation areas" of plastic metals with stage destruction. In our case, when the thermo-dynamic material condition is changed, it is possible that some structural formations of PE decrease their role with destruction, while the others become crucially important. It is due not only to macroscopic character of quasi-ductile failure and the decrease of deformability, but to the local crystallization of the material in the area of plastic deformation development start.



**STUDY OF INFLUENCE OF DISPERSIBLE CARBON ON
ELECTROMECHANICAL PROPERTIES OF
ELECTROCONDUCTING MATERIALS ON THE BASED ON
ALKALINE BINDER SYSTEMS**

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It's determined the possibility of deriving for nonlinear resistances based on system " β -C₂S-SiC-Na₂OnSiO₂mH₂O" described by compressive strength 38MPa, resistivity $\rho=4.05 \cdot 10^5$ Ohm·m and nonlinear coefficient $\alpha=0.18$. These materials are using as varistors in electrical systems of control.

For expansion of range application of received materials, including callefactors, the possibility of modification or system " β -C₂S-SiC-Na₂OnSiO₂mH₂O" by the component of a powder of graphite is surveyed.

According to data [1,2], the introduction of dispersible carbon in composition of resistive aggregates including silicon carbide, allows to reduce values of contact resistances between grains of silicon carbide and to receive common 'working' resistance of compositional material in limits 0.1÷0.3 Ohm·m. Also it is known [3], that support in composition of electroconducting concrete (betels) the carbon component results in a drop of his strength and durability.

With the purpose of deriving resistors, capable to work in a condition of long-lived insert distinguished resistivity in limits (0,1÷ 0,3) Ohm·m and compressive strength not less than 7 MPa, the optimization of composition of the above-stated materials with usage a simplex - lattice method of experiment is realized [4].

As starting builders have utilized tehnogenic raw material contented of β -C₂S ($X_1=10\text{--}40\%$), water glass with silicate modulus $M_s=2,0$ and density 1250 kg/m³), powder of silicon carbide ($X_2=55\text{--}80\%$). For modification of compositions have utilized dispersible carbon ($X_3=5\text{--}35\%$), ground up to specific surface 750 m²/kg. Samples (diameter and height are equal 22 mm) gained by a method of a cold molding ($P_{compr}=50$ MPa and dampness of fusion mixture 8 %), exposed drying at temperature 373K and burning at temperatures 773K and 1273K.



The isoparametric diagrams of changes in compressive strength and resistivity are build in the case of mathematical haudind of recived date (fig.).

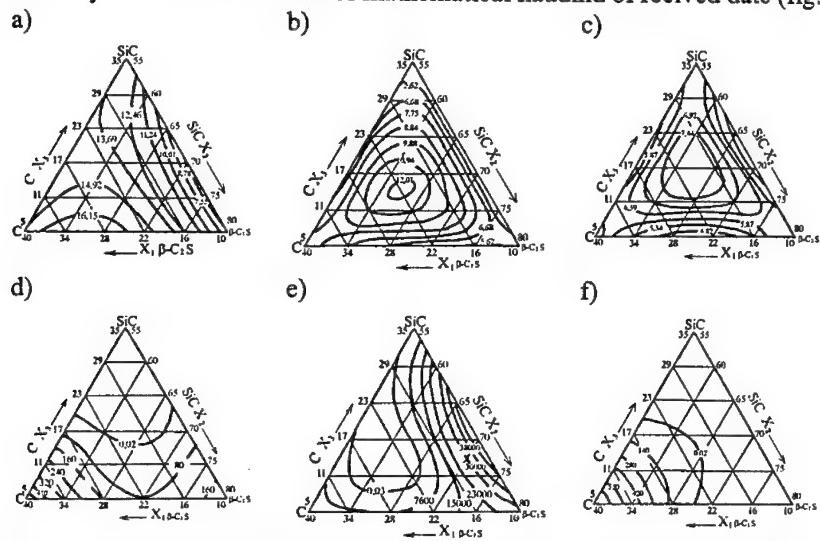


Fig. The isoparametric diagrams of changes in compressive strength (a, b, c) and resistivities for resistive compositions (d, e, f) after heat treatment at temperatures 373K (a, d), 773K (b, e) and 1273K (c, f).

The analysis of reduced data displays, that the reaching of operating range of resistances is supervised in the field of compositions including the component of graphite in an amount from 10 up to 16 %, thus the drop of compressive strength after burning at temperature 1273K constitutes 53%.

The received materials ($R_{\text{compr}}=7.5 \text{ MPa}$ and $\rho=0.2 \text{ Ohm}\cdot\text{m}$) on the operational properties meet the requirements, which need for to heating elements on the basis of mineral binders and can be recommended for usage as heating elements of thermal devices both household and industrial assigning.

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THE INFLUENCE OF THE ORIENTATIONAL RELAXATION OF THE MOLECULES ON THE MOTION OF DISLOCATIONS IN C₆₀ FULLERITE

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The study of plastic deformation of single crystals of C₆₀ fullerite have shown that over a wide range of moderately low temperatures 300-80 K the carriers of plasticity in them are dislocations belonging to the slip system (111)<110>. Therefore, in constructing a consistent microscopic theory of the plasticity and strength of crystalline C₆₀, one of the first problems is to analyze the mobility of dislocations in this material.

In the low-temperature simple cubic (sc) phase of C₆₀ fullerite, which exists below T_c=260 K, the most efficient mechanism of dissipation of mechanical energy is the lattice-orientation interaction and the orientational relaxation of molecules – thermally activated transitions in the double-well potential between two energetically inequivalent orientations of molecules, which have come to be called the pentagonal (p) and hexagonal (h) configurations. The p configuration is more energy expedient and it is separated from h configuration by an energy barrier U_p≈0,3 eB, and the energy of the h configuration is U_h=U_p-Δ, where Δ≈0,01 eB. On the basis of energy consideration let us suppose that at T→0 the ideal thermodynamic equilibrium structure of the fullerite corresponds to the p configuration of all molecules, and the h configuration should be regarded as local structural defects that can be excited, for example, by the thermal motion of the molecules or the deformation of the crystal lattice. The volume densities N_p and N_h of the configurations should satisfy the balance relation

$$N_p + N_h = N_0 \quad (1)$$

where N₀ is the volume concentration of the double-well orientational states. To describe the relation of the rotational degrees of freedom of the molecules with the long-wave deformations of the crystal lattice, it is necessary to introduce deformation corrections to the parameters U_p, U_h, and Δ:

$$U_{p,h}^{(\varepsilon)} = U_{p,h} - v_{p,h} \varepsilon_{||}(\vec{r}, t) \quad \Delta^{(\varepsilon)} = \Delta - v_\Delta \varepsilon_{||}(\vec{r}, t) \quad (2)$$

where ε_{||} is the sum of the diagonal components of the strain tensor, and v_p, v_h, and v_Δ=v_p-v_h are the constants of the deformation potential. Let us assume that the orientational relaxation (thermally activated transitions between the p and h configurations) occurs over time interval

$$\tau_{p,h}^{(\varepsilon)} = \tau_0 \exp \left(\frac{U_{p,h}^{(\varepsilon)}}{kT} \right), \quad (3)$$



where τ_0 is the characteristic librational period.

The stated above concepts and relations (2) - (3) let us describe the orientation-dislocation interaction and the influence of the interaction on the motion of dislocations if we consider the dilatation field around the single dislocation line as a deformation ε_{\parallel} . The presence of a dislocation in a C₆₀ fullerite crystal disrupts the uniform distribution of pentagonal and hexagonal configurations of the molecules. A nonuniform equilibrium "atmosphere" of defect h configurations, similar to the Snoek atmosphere forms in the elastic field ε_{\parallel} of a sessile dislocation, and the starting force F_s(T) exist that is capable of tearing the dislocation away from the cloud of h configurations. For the edge dislocation this force is equal:

$$F_s(T) = \left(\frac{3G}{3B+4G} \right)^2 \frac{b^2 v_{\Delta}^2 N_0}{4\pi r_0 kT} \frac{\exp(\Delta/kT)}{[1+\exp(\Delta/kT)]^2} \quad (4)$$

Here b and r₀ are Burgers vector and the radius of the dislocation core, B and G are bulk modulus and shear modulus.

The strain field $\varepsilon_{ik}(\vec{r} - \vec{V}t)$ of a dislocation moving at a constant velocity V disrupts the local thermodynamic equilibrium between the p and h configurations and excites a relaxation process that restores this equilibrium. On account of the energy dissipation accompanying this relaxation process the dislocation experiences an equivalent dynamic drag force F_D(T,V):

$$F_D(T,V) = F_s(T) I(V,T) \quad (5)$$

Here I(V,T) is the complicated function of V and T, but I(V,T) ≤ 1 for all values of these arguments, therefore the characteristic scale of the force of orientational drag on dislocations is set by the starting force F_s(T). A characteristic feature of the dynamic component of the drag force is that its dependence on the velocity V has a broad peak with an almost flat top, and the position of this peak on the velocity axis depends exponentially on the temperature. Formulas (4) and (5) let us make a qualitative interpretation of the features on the temperature dependence of C₆₀ fullerite microhardness in the low temperature region detected in the experiments.



SLIP SYSTEMS AND DISLOCATIONS IN C₆₀ CRYSTALS

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Fullerite C₆₀ is a typical simple molecular crystal in which peculiarities of lattice properties such as polymorphism are associated with the thermal activity of rotational degrees of freedom of molecules. The phase transition observed at $T_c = 260$ K (which is a first-order phase transition according to some indications) does not change the symmetry of spatial arrangement of the centers of gravity of molecules that form the fcc lattice. A decrease in lattice symmetry from Fm3m (fcc) to Pa3 (sc) upon cooling is due to partial orientational ordering of molecules. The molecules in the low-temperature phase are in two orientational states that are non-equivalent from the symmetry point of view, but have close energy values. These states are called pentagonal and hexagonal configurations and correspond to global and local energy minima of the noncentral part of molecular interaction.

It has been established that plastic deformation of C₆₀ crystals has a dislocation nature. Optical microscopy revealed a well-developed system of slip lines in the vicinity of the indentation [1,2], while thermal and chemical etching revealed typical systems of dislocation etch patterns [3]. It could be expected from general considerations that an fcc \rightarrow sc phase transition can lead to the replacement of the easy slip system <110>{111} by another system typical of the sc lattice (e.g., by <100>{100} or <110>{100}).

The high-temperature fcc phase of fullerite, which is dominated by central van der Waals molecular interaction, can be formally treated as a "plastic crystal". A transition to the orientationally ordered SC phase is accompanied by an enhancement of the role off-central component in molecular interaction, and must increase the lattice rigidity. Hence, in addition to slip geometry study, investigations of the temperature dependence of the parameters of plastic deformation of C₆₀ crystals especially in the vicinity of the fcc \rightarrow sc phase transition also acquire a considerable significance.

Static and dynamic indentation technique are used for mechanical testing in most cases because of the small size crystals being grown. In this report, the authors' results on the study of slip systems and micromechanical properties of C₆₀ crystals received from gas phase are given.

The pattern of slip traces in deformed regions in the vicinity of indentations on two habitus planes (111) and (100). Crystallographic analysis of the patterns of plastic shears lead to the unambiguous conclusion concerning the activity of the only slip system of the {111}<110> type in the fcc and sc phases. The conservation of slip in {111}-type planes during fcc \rightarrow sc phase transition is a phenomenon that has been studied well in ordered alloys Cu₃Au and Ni₃Mn. It is pos-



sible that like in these alloys, slip in {111} planes in the ordered sc-phase of C₆₀ occurs through the motion of partial dislocations accompanied with the formation of low-energy antiphase boundaries.

The temperature dependence of microhardness H_V of C₆₀ crystals is obtained in the temperature interval 77-300 K. At room temperature, the typical value of H_V ≈ 0.2 GPa. As regards microhardness, fullerite C₆₀ crystals are comparable to graphite, plastic fcc metals of the gold-type, or NaCl crystals. Weak van der Waals interactions between molecules are responsible for low values of microhardness. The hardness of the fcc phase normalized to the elastic modulus turns out to be higher than the hardness of typical molecular crystals at comparable homological temperature values. This fact as well as considerable localization of slip indicates that the crystals studied cannot be classified as plastic crystals in view of the presence of hardening impurities in them.

Plastic deformation in C₆₀ crystals is anisotropic, which is manifested, for example, in the dependence of the value of H_V on the direction of the indentation plane: H_V⁽¹¹¹⁾ = (1.25 - 1.5)H_V⁽¹⁰⁰⁾ at T = 300 K.

The following features are revealed on the temperature dependence of microhardness. The microhardness increases stepwise by approximately 30 % upon a transition through T_c, while at T < 160 K the H_V(T) dependence becomes much stronger. The increase in the value of H_V during the fcc → sc transition is observed in the temperature interval large than 10 K. The width of this interval decreases as a result of annealing in vacuum. These anomalies are regarded as consequences of dislocation drag due to relaxation losses during the interaction of the elastic field of dislocations with the field of the orientational order parameter (the T_c region) and with the system of pentagonal and hexagonal configurations of C₆₀ molecules whose equilibrium is violated by a moving dislocation (the region of T ≈ 160 K).

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HARDENING OF HIGH-NITROGEN STAINLESS STEELS

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Austenitic stainless steels alloyed with nitrogen belong to a new class of prospective structured materials possessing high strength and resistance to corrosion. Deformation under high hydrostatic pressure (HHP) is one of the effective methods for the formation of the optimal defect structure of materials, which provides a high level of mechanical properties and operating characteristics. It is known that under the hydrostatic extrusion a material is more uniformly processed over the whole of its volume. High-pressure liquid used as a deforming tool enables one to realize high single deformations in the absence of material failure and produce articles of better quality of the surface [1]. It is also advisable to use HHP with austenitic steels since it suppresses transformations accompanied by the increase in volume, thus preventing deformation-martensite formation, which, in turn, makes it possible to produce articles in nonmagnetic state.

In the paper, structure changes and the character of deformation hardening of austenitic steels of system Fe-Cr-Mn-Ni with the nitrogen content approximately equal to 0,7 wt% have been studied after the stressed-strained state action by different schemes and the thermal treatment. Austenitic steels X19Г10ФС2 and X22Г15Н8ФМФ have been investigated.

The austenitic steel strength can be improved by solid-solution alloying, plastic deformation, ageing. The alloying of steels by nitrogen results in the abrupt strengthening. The age-hardening also depends on the nitrogen content of steel, however, it is expedient to use this method of strengthening the above steels with the nitrogen content of 0,4wt% and higher.

The nitrogen, which dissolves in austenite interstitially, promotes the high hardening of austenitic alloys under the cold plastic deformation. This is due to a noticeable decrease in the level of stacking-fault energy, which to a considerable degree characterizes the complex of physical and mechanical properties of materials. The effect of hardening under deformation grows with the increase of nitrogen concentration in steel and of deformation degree.

The behavior of the nitrogen-containing austenitic steels depending on scheme of the stressed-strained state, has been studied. It is shown that the highest complex of the mechanical properties was obtained after deformation with hydrostatic extrusion, not in the result of drawing or uniaxial compression.

Due to the joint action of nitrogen and HHP a disperse twin structure is formed under the deformation providing a high complex of mechanical properties in the investigated steels. The electron-microscope investigations have shown that HHP makes the process of deformation twinning easier, favour an earlier and



intense twinning as compared to deformation under the atmospheric pressure. The intensity of twinning increases also with the growth of deformation degree. When twins from different twinning systems are intersecting they crush and the structure becomes of high dispersivity [2].

The nitrogen in the form of nitrides causes the dispersion hardening which additionally improves the strength characteristics of steels belonging to this class. The efficiency of hardening under the age hardening is specified by the correct selection of hardening-phase type, quantity of the phase, its distribution, it grows with the increase of nitrogen concentration in steel.

The formation of the disperse twin structure during the deformation under high pressures and subsequent decomposition of the solid solution enable one to obtain a high complex of mechanical properties of the investigated steels in the nonmagnetic state: $\sigma_B=2050$ MPa, $\sigma_{0,2}=1950$ MPa, $\varphi=33\%$ $\delta=7.8\%$.

Thus, the obtained experimental results are the basis in developing the optimal methods of the austenitic-alloy hardening under the deformation and thermal treatment. The hardening of high-nitrogen steels based on solid solution of Fe-Cr-Mn-Ni by means of hydroextrusion followed by ageing is a prospective method for the purposeful obtaining of high-level mechanical properties with steels remained nonmagnetic.

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ON TOUGHNESS OF CONCRETE UNDER HIGH TEMPERATURES

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Known that when increasing a temperature occurs considerable reduction of toughness of all concretes [1]. This is stipulated not only structure changes to the concrete, connected with the physicist-chemical reactions (evaporation adsorbed moisture and liberation crystal moisture) [2], but as well as "breaking a crystalline structure cement stone in the concrete" [1]. Because of differences a physicist-chemical characteristics cement stone and fill in materials at the heating occurs a weakening of adhesive powers in contact layer between these main components of concrete. Before this time a level appearing in the concrete of voltages at the heating before high (near 700 K) temperatures specially was not researched. The purpose given paper consisted in numeral analysis of floors of voltages, appearing in ben-not at the heating before high temperatures on the base models, taking structure spottiness of concrete into account.

Speech a problem on the tense condition of square-wave band concrete, warmed before high temperatures. At the statement of the problem considered concrete by the porous ambience with partly filled moisture times. Moisture are absorbed from atmosphere. Heating of concrete is realized on one of the borders of square-wave area of deciding a problem to the account convective and radiational heat exchange with the external ambience. When raising a temperature begins process of evaporation moisture, being present in microtimes. With the growing temperature a process of evaporation are intensificated and grows a pressure of products of evaporation in times. Under nonwhich swing of pressure on the thickness layer begins a filtration of vapours of water to warm surfaces, intensification which depends on gas penetrating concrete. At the achievement temperature beginning of process an dehydrotation begins a process of liberation crystalline moisture. To the account additional gas ceccretion grows pressure in microtimes and voltage level. Including prototyped fill in materials concrete. Calculation of warm-up voltages was conducted within the framework of model flat tense condition [3].

Worded problem on the tense condition of square-wave band of concrete with including a speech by the method of end differences. Numeral analogues of differential nonlinear equations a speech by the method the iterations with using a method of proracing on each iterations [4].

Numeral realization worded numeral problem has a number of particularities, rolling front of evaporation stipulated by presence of and areas an dehydrotation, in which is realized intensive local heat absorption. Presence of front of evaporation brings about need used of irregular and noneven numeral net,



which parameters were chosen from conditions of ensuring stability and convergence of numeral decision. Realignment of net was conducted on each step on a time - inicken in vicinities of front of absorbing a heat and discharging on lengths, through which front of evaporation pass. Values of sought functions in nodes of new net were defined by the interpolation on values of functions in nodes of net with preceding step on a time. Total number of nodes reached herewith 5.104.

The results numeral analysis have shown that appearing in the concrete at the heating before high (near 700 K) temperatures of voltage exceed limit toughness typical (including and highmark) concretes under corresponding temperatures. Installed, in particular that on the value arising voltages affect not only features of cement stone, as well as fill in materials. Characteristic of the concrete is presence greater gradient pressure on spatial coordinates. On the measure heat concrete and transition adsorbed and crystalline moisture in vaporous co-costing an area, in which pays a maximum of pressure, is displaced from warming surface into material.

Was it specially researched influence source ("initial") porous of concrete and contents adsorbed moisture on the value arising voltages. Numeral studies have shown that upmark concretes, having comparatively high (in contrast with highmark concretes) source porosity, better withstand an influence both high temperatures, and flows of vapours of water, standing out as a result dehydrotation concrete. In ditto time high contents adsorbed moisture in times of upmark concrete brings about more high gradients of pressure of vapours and accordingly to raising a level internal strain in the concrete.

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ON DESTROYING STEEL AT INTERACTION BY HIGHETEMPERATURE HETEROGENEOUS TREEMS

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Installed [1] that destruction of steel at the interaction with super-sound velocity highwarm-up heterogeneous streems, containing particles of aluminum and ferric oxide and attacking on the barrier under the direct angle, occurs thermomechanical destruction in the mode of. In this mode a temperature of destroying surface does not exceed 1000 K at velocities of destroying before $4 \cdot 10^{-2}$ m/s, but heterogeneous flow acts upon the material as an abrasive, deforming fine under-surface layer, heated before the tempera-aurochs, under which toughness of material repeatedly decreases. Hard or fluid particles presensing in the strem reveals itself in intensification a pro-cession an heat exchange streems with the material and in increase tangent strain of friction of flow to the account of growing of efficient density of mixture "gas - particles". Destruction steels occurs in conditions intensive heat gas-dynamic influence of highwarm-up flow, if values of tangent voltages of friction τ exceed a limit of toughness σ_n steels under the correspond temperature. With the growing of temperature of value s_{II} all steel fall. Velocity of this reducing, on the measure of increasing a temperature, races-aunt [2,3]. Particles, moving on paths, directed on tangent to surfaces, act upon the shallow metal layer, heating before high temperatures, in the same way either as molecules of gas - in the mode "friction-slides". As a result fine under-surface layer of material go to ruin on the depth, in which is met the condition $\tau = \sigma_n$.

In given work a speech a problem of calculation of tangent voltages of friction τ heterogeous flow, equal limit of toughness destroying matterials on the sprain or cut under the warm-up field under-surface layer, meeting the terms heat exchange material with highwarm-up heterogeneous strem.

From deciding a problem of heat conduction with the rolling border, which velocity of motion was experimented to find, paid a temperature fields in the material and field of limits of toughness σ_n , corresponding values of temperature in each spot of area of decision. Were they Then defined values s_{II} on surfaces of destroying, equal τ .

The results of numeral analysis have shown that level of voltages of friction homogeneous flow of vastly below values σ_n , corresponding installed in the experiment to values a velocity of destroying typical steel. Values τ



heterogeneous flows, hanging from concentration of hard or fluid particles, reach values required for destroying heated before high temperatures (but not melted) under-shallow layer of material. On the grounds of collations of results numeral analysis and experimental given made conclusion on possibility realization of steel mechanism of thermomechanical destroying in exactly broad range of change as gasthermodynamics parameters of heterogeneous streems, so and physicst-mechanical features of steel.

Special numeral analysis is conducted for the reason separations of range of changing the concentrations of hard particles in near-wall frontier layer, in which are reached condition of destroying a surface of metal. Considered different mass concentrations of hard phase in the streem and hydrodynamic modes of attack streem on the barrier. Installed that under mass concentration from 0,2 and more in near-wall area is formed utter layer particles, interacting with destroying surface in the mode "rub-thread-slides". At thickness layer can reach two distinctive sizes of particles, but three-dementional concentration of particles in frontier layer 0,2. Such level to concentrations of hard phase in near-wall area corresponded values of voltages of friction, greatly exceeding limits firmly typical carbon steel at temperatures 800 - 900 K.

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NUMERICAL FORECASTING OF CONNECTION STRENGTH OF COVERS WITH THE BASIS AT PLASMA POWDER

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Plasma powder of covers is one from perspective process engineerings both modernity, and future [1]. But this process engineering requires at a modern level of a realization large heat expenditure, as a rule, usual industrial firms, incompatible to capabilities. Therefore the further promoting in the market scientific capacious process engineerings is connected to minimization technological heat expenditure. The solution of this problem by large-scale full-scale experiments is represented improbable. It is expedient to conduct selection of the technological schemes, modes, parameters with reference to the given problem by computing experiments on the basis of mathematical models circumscribing researched process with the greatest possible completeness.

The purpose of the given activity was the forecasting of strength of connection of covers with the basis at plasma powder on the basis of main rules of model [2], taking into account non-permanent of a temperature field in a zone of a contact of a particle of powder of a material and details. The relation of strength of connection powder of a material with the basis from a complex of the factors influencing to size of temperature of a zone of a contact T_k of a single particle with a surface of a details was investigated, on which the cover is put.

The numerical analysis of regularities of chemical interaction of a particle with the smooth basis conducted at use of assumption about equivalent energization of all atoms of a particle from a leg of a details. In this case speed of a chemical response is determined only by conditions of activation of atoms of a surface of a fundamentals. The problem about a temperature field of a system " a particle - basis " for a non-stationary mode of interaction of a plasma jet with a details is decided. The relation of temperature of a zone of a contact to time was used as the input characteristic at calculation of a degree of reaching of maximum strength of coupling of a particle with the basis.

The speed of topochemical reaction was calculated with use of a differential equation [1].

$$\frac{dN(t)}{dt} = [N_0 - N(t)] \sqrt{\exp\left(-\frac{E}{kT_k}\right)}$$

Here: N_0 - number of atoms on a surface of a fundamentals or particle located in a mutual physical contact; $N(t)$ - quantity of atoms from number N_0 reacting during



t; E - energy of activation; n - frequency of own oscillations of atoms; k - constant by Boltzman.

The relative strength of coupling of particles with the basis thus is determined by a ratio:

$$\varepsilon = \frac{\sigma(t)}{\sigma_{\max}} = \frac{N(t)}{N_0},$$

where $\sigma(t)$ - strength achieved during t, σ_{\max} - maximum strength which reach till completion of process.

The formulated thus mathematical model describes communication of the integral characteristic e with parameters of a thermal mode technological powder of cover.

The problem is decided by a method of final differences. For the solution of two-dimensional equations the method was used locally - one-dimensional. At the solution of one-dimensional equations the implicit four-dot iterative scheme was applied. The grid parameters were selected so that for a particle of any size to supply not less than 15 units on each coordinate. The irregular and irregular incremental grid was used.

The main calculations are executed at the following values of parameters of external environment and characteristics of materials: temperature of plasma $T_{\Pi} = 3500K$; a factor of heat exchange $\alpha = 10^3 W/m^2 K$; energy of activation . $E = 3,236 \cdot 10^{-19} J$; frequency of own oscillations of atoms $v = 10^{13} c^{-1}$; a constant by Stefan- Boltzman $k = 1.38 \cdot 10^{-23} J/K$; the sizes of a particle $d = 1 \cdot 10^{-4} m$; a reduced power of blackness of plasma $\varepsilon_n = 0.5$; heat conduction of a material of a particle and fundamentals $\lambda = 46 W/mK$; thermal capacity of a material of a particle and fundamentals $c = 500 J/kgK$; density $\rho = 7800 kg/m^3$; reference temperature of a fundamentals $T_0 = 1500 K$.

Is established(installed), that the size T_k essentially depends on time. Accordingly and the sold strength of coupling is determined largely by kind of this relation. Is established also, that mathematical model [2] allows to conduct evaluations of parameters of technological process powder in enough broad band of change of the environmental conditions.

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COMPUTER SIMULATION OF FRACTURE PROCESS OF HETEROGENEOUS MATERIALS UNDER SHEAR LOAD

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A 2D computer model for fracture of heterogeneous material with block multi-level structure was developed. This model relies on the kinetic concept of strength of solids [1]. The model allows us to study the fracture development not only in space, but also in time, under different conditions of loading. The model gives the possibility to see how properties of a material affect the fracture process.

The model material is a two-level system of elements. The lower level is formed by structural elements equal in size. The upper level is formed by blocks of adjacent elements from the lower level. The blocks are different in size (i.e. in the number of the elements), shape, and strength. Boundaries between the blocks and between the elements are described by different functions of the stress redistribution.

The computer experiments simulate the material behaviour under the shear stress. The lifetime of each element is determined on the assumption that the most probable time from load application to breakdown of the element exponentially depends on stress

$$\tau_i = A \exp(-\gamma_i \cdot \sigma_i), \quad (1)$$

where γ_i is the structure-sensitive parameter and σ_i is the stress on the i -th element and A is a constant. The heterogeneity of the system is modeled by assigning a parameter γ from a given distribution to each element. In order to take into account the random character of fracture, it is set the probability of fracture of the i -th element:

$$P_i(t_i) = 1 - \exp(-t_i / \tau_i), \quad (2)$$

where t_i is the actual time before the fracture of the i -th element which differs from the most probable time (τ_i). Then the random time before the fracture of an element t_i is given by the relations:

$$\int_0^t \frac{dt}{A \exp(-\gamma_i \cdot \sigma_i)} = \ln\left(\frac{1}{1 - P_i}\right). \quad (3)$$

In the computer experiment, lifetime of all intact elements are calculated. The element with the minimum lifetime is regarded as fractured. When the i -th element is destroyed, the stress from it is redistributed to nearest neighbors in radius R (parameter of the model).



Computer experiment finishes when the last element fractured.

The result of computer experiment is a series of defects, which are characterized by nucleation time, coordinates and size. Fracture development in time and place is analyzed by statistical methods. Simulation results are comparing with experimental data, which is received at granite samples fracture [2]. It is shown, that fracture regularities (evolution of fracture nucleation site), which are found in experiments, are also observed in the simulation. Furthermore, the model has an ability of stick-slip researching, this phenomena is characteristic feature of rock fracture. Partially fractured surface is simulating for imitation of fault, in which area stick-slip is happened. Methods of assignment of different levels of stress on elements in fault area, where stick-slip happens, are studying. Model parameters are determined by comparing with experiment of deformation of granite sample with natural made fault.

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THE RESEARCH OF THE DYNAMICS OF CRACK DEVELOPMENT IN CRYSTAL LATTICE OF Al

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The research of structure-energetical transformation of initially ideal crystal under the influence of different types of deformation from the elastic stage to full fracture were made by the method of variational quasistatic in [1-3]. The FCC crystal of solid Ar was chosen as the object of the research. It was built in quasithreedimensional approximation from interacting atomic chains in approximation of pair interatomic Lennard-Jones potential in the direction <211>. Computer experiments showed, that deforming crystal passed the set of stages of structural reconstruction at the loading. Every stage had definite types of defects in the structure: displacement of atoms, formation of dislocations, nucleation of dislocations, formation of pores and cracks.

It was shown [4], that orientation and form of crack changed from zigzag in the direction <110> to stepped form, and again to zigzag form in the direction <112> in the case of combined deforming influence on the crystal (tension in the direction <112> and shears along <110> in the dependence on relation of deforming tension). When the crack-initiator was inculcated in the structure of the crystal in the direction, perpendicular to tension deformation, the value of maximum deforming tension, leading to fracture, decreased. The level of tension decreasing of full fracture was inverse proportional to the length of the crack-initiator.

It was made the research of the dynamics of the crack development on the example of FCC crystal of Al in the present paper. The influence of temperature and tension on the velocity of the crack development, including 4000 atoms in the plane (111), was investigated by the method of molecular dynamics. Free boundary conditions were applied outside the limits of the calculated block. Interactions between atoms were given in the approximation of Morse potentials.

The crack velocity was found under different ranges of temperature (from 300-350 K). The crack development was perpendicular to tension stress in the direction [111] and the velocity of the crack growth was measured at each 0.45 ps. It was observed, that the crack propagated in ductile manner under different ranges of temperature and the velocity of the crack growth was unstable. Figure 1 shows the simulation result under strain strength 50 mN/m and at temperature 310 K. We suppose that the crack can reach a point of dynamic instability at the velocity less than the Rayleigh wave velocity.



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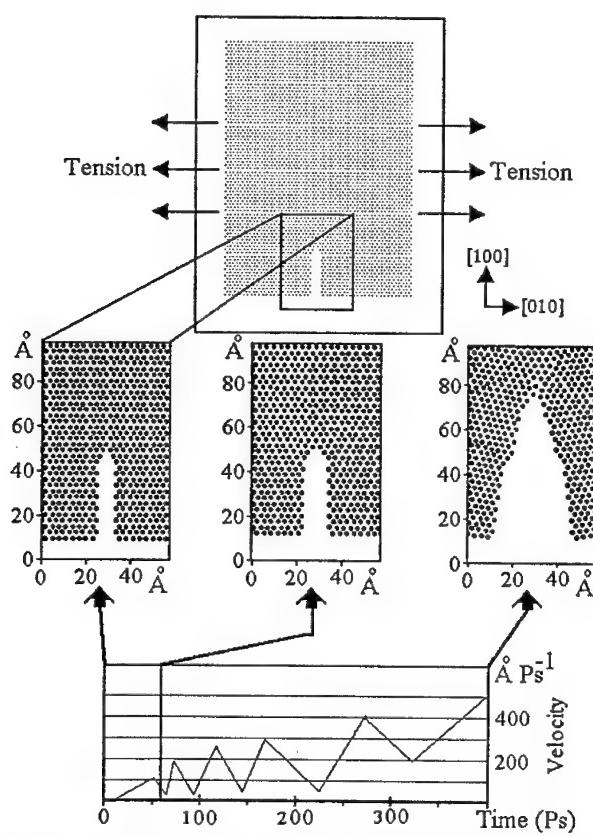


Fig.1. The crack growth under tension 50 mN/m and at 310 K.



**THE INVESTIGATIONS OF STRENGTH CHARACTERISTICS OF
COMPOSITE MATERIALS BASED ON THERMOEXFOLIATED
GRAPHITE**

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The strength of the samples of carbon graphite materials was estimated from experimentally measured $\sigma(\epsilon)$ dependences obtained in conditions of continuously growing mechanical strain σ in the range $\sigma \in [0, \sigma_D]$ and increasing relative elongation $\epsilon \in [0, \epsilon_D]$, where σ_D and ϵ_D are the limit of mechanical load and the limiting relative elongation respectively, which are responsible for sample's destruction.

The samples that have been investigated are composite materials based on TEG with different types of binders (polyvinyl acetate (PVA), phenol formaldehyde resin (PFR), silicon organic compound (SO)). The results of the investigations are given in Table 1.

Table 1.

Sample	ρ , g/cm ³	TEG, % mas.	σ_D , MPa	ϵ_D	$\left(\frac{\partial \sigma}{\partial \epsilon} \right)_{MAX}$
TEG	1.91		40.0	0.168	209
TEG	1.94		20.0	0.08	240
TEG+PVA	1.414	21	40.0	0.26	285
TEG+PFR	1.290	58	36.0	0.18	237
TPГ+ФФС	0.93	26.7	28.2	0.425	208
TPГ+ФФС	0.896	26.7	24.2	0.391	218

It was found that addition of PVA to TEG does not lead to the change of the sample's strength while PFR used as a binder in CM makes the strength of the CM smaller, and the larger the PFR concentration, the less strength the CM displays under compression.

In the samples of "TEG+SO" the process of mechanical destruction when the sample cracks up into separate pieces does not occur at all: when squeezed the sample went through the gap between the press punches. Under initial small loads $\sigma(\epsilon)$ dependence is close to the linear one, which is characteristic of composite materials based on TEG and which is the result of transformation of the sample's



characteristics in the process of mechanical deformation. With further growth of the load (and relative elongation respectively) the slope $\left(\frac{d\sigma}{d\varepsilon}\right)$ on $\sigma(\varepsilon)$ dependence becomes less steep and at a certain value of relative elongation $\sigma(\varepsilon)$ dependence reaches the maximum. However, in contrast to composite materials investigated previously ("TEG+PVA", "TEG+PFR", pure compacted TEG) this maximum is local and is not accompanied with sample's crumbling. Instead the intensive viscous squeeze out is observed when the material escapes through the gap between the press punches. As a result $\sigma(\varepsilon)$ dependence shows a local minimum after which $\sigma(\varepsilon)$ begins to grow gradually.

The analysis of the results obtained for composite materials "TEG+PFR+CF" showed that $\sigma(\varepsilon)$ dependence, which is a function of mechanical load σ on relative elongation ε , is non-monotonous for the samples of "TEG+PFR+CF" with 60% mass of TEG and 1.7 % of carbon fiber. This is likely to be due to that the development of macroscopic cracks in the sample is slowed down to some extent by carbon fiber that acts as a stopper on the way of developing crack.

The investigations of creeping processes in composite materials "TEG+SO" and "TEG+PFR+CF" showed that the behavior of $\varepsilon(t; \sigma)$ dependence is typical for CM of given composition and agrees well with the results of investigations in their mechanical properties. It has been found that CM based TEG-organic compound belong to the class of viscoelastic materials which distinguishes them from CM based on pure TEG which belong to the class of elastoplastic materials.

The low value of instantaneous elasticity modulus and intensive creeping process under constant loading is characteristic of composite materials "TEG+SO". The direct creepage prevails essentially over the reverse one and the sample does not recover in size even after exposure without load for a day or two. The creepage curves for "TEG+PFR" and "TEG+PFR+CF" differ greatly from those obtained for "TEG+SO".

It was found that composite materials "TEG+PFR+CF" are more elastic than "TEG+PFR". So, the hop-like changes of relative elongation observed for "TEG+PFR+CF" with the change of mechanical load are greater than the analogous changes for "TEG+PFR" and the contribution of viscosity component into $\varepsilon(t; \sigma)$ is smaller for "TEG+PFR+CF" than for "TEG+PFR".



**THE INVESTIGATION OF ELASTIC CHARACTERISTICS OF
COMPOSITE MATERIALS BASED ON THERMOEXFOLIATED
GRAPHITE**

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With the aim to determine the elastic and plastic characteristics of composite materials based on TEG with different types of binders (phenol-formaldehyde resin (PFR), silicon organic compound (SO)) the "loading-unloading" diagrams obtained experimentally at room temperature and at the temperature 473 K were analyzed and the efficient elasticity module E_{eff} as well as maximal deformation ϵ_{max} and residual relative deformation ϵ_{res} were determined. The obtained data are presented in Table 1.

Table 1.
Mechanical characteristics of composite materials based on TEG.

Material	T, K	Cycle	E_{eff} , MPa	ϵ_{max}	ϵ_{res}
"TEG+SO"	293	1	105 ± 20	0.2 ± 0.02	0.1 ± 0.03
"TEG+SO"	293	2	430 ± 67	0.02 ± 0.007	0.02 ± 0.01
"TEG+PFR"	473	1	80 ± 17	0.15 ± 0.03	0.1 ± 0.01
"TEG+PFR"	473	2	250 ± 90	0.05 ± 0.03	0
"TEG+PFR+ +CF" (42%)	293	1	45 ± 7	0.25 ± 0.03	0.11 ± 0.01
"TEG+PFR+ +CF" (42%)	293	2	68 ± 14	0.17 ± 0.03	0
"TEG+PFR+ +CF" (60%)	293	1	130 ± 24	0.08 ± 0.01	0.04 ± 0.01
"TEG+PFR+ +CF" (60%)	293	2	226 ± 11	0.05 ± 0.002	0

Since the properties of TEG-based CM are known to change essentially within the first compression and to get stabilized after the first unloading they were investigated within two or three "loading-unloading" cycles. All the three cycles are not closed, i.e. the residual strain may arise in any of them and its value is comparable with the diagram's half-width for the given cycle.

It was found that the efficient elasticity module tends to grow within the first cycle as the content of TEG in the CM is increased while within the second and the third "loading-unloading" cycles this tendency was not observed. This is



probably due to the fact that the mechanism of plastic deformation depends essentially on the concentration of composite material components.

At the temperature of 473K "loading-unloading" cycling proved to be completely impossible because even at small mechanical loads applied to them the samples were squeezed out through the gap between the punches. After the complete removal of load the samples did not show even the smallest resemblance to their initial shape.

Thus, at the increased temperatures the samples of composite materials "TEG+SO" with different concentrations of TEG behave like viscous liquids. Therefore it is unreasonable to use the composite materials of this kind for producing the work pieces that have to endure considerable pressures in conditions of high temperature (for example, high temperature sealings).

The investigation of the samples of composite materials "TEG+PFR" that were subjected to cold pressing followed by the exposure at the temperature 473 K for polymerizing phenol formaldehyde resin were performed. The "loading-unloading" diagrams obtained at the temperature 293 and 473K are similar in their general form, and their numerical characteristics are rather close, which testifies to high thermal stability of mechanical characteristics of "TEG+PFR" CM.

The investigations of CM "TEG+phenol formaldehyde resin+carbon fiber" (TEG+PFR+CF) showed that carbon fiber added to the CM even in small concentration makes the efficient elasticity module and maximal relative deformation essentially more stable, that is the mechanical properties of the CM that contains carbon fiber are transformed much less within the first "loading unloading" cycle than those of the CM without carbon fiber.



MECHANICAL PROPERTIES OF THE FINE GRAIN FE-CU PSEUDOALLOYS

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The mechanical properties and microstructures of the fine grain (the grain sizes are less than 1 μm) Fe-30%(27,3%vol) Cu pseudoalloys were investigated. The pseudoalloys were heated in the solid state up to 1000°C and above the copper melting temperature (1130°C).

Some of the samples with relative density more than 96% were prepared by sintering of the fine size powder iron-copper blends in solid state, then were deformed and annealed for the residual porosity was minimum. The other samples were prepared by the liquid state sintering. The resistivity (ρ), hardness (HB) and the mechanical properties during tension were determined. The computer image analyzer "SIAMS" was used for the quantitative analysis of microstructures.

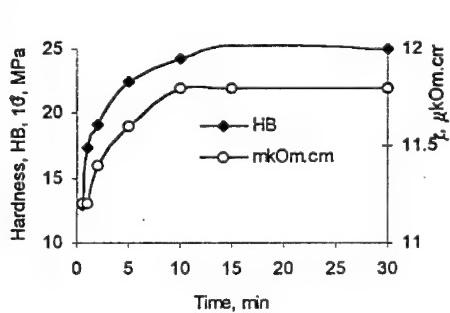


Fig.1. The hardness and resistivity of pseudoalloys as a function of annealing time at 1130°C

stabilized after 10 min of sintering (fig.1). The grain sizes decrease when the densification is over, then the grain sizes increase as the exposure increases too. The mechanism of the grain growth is the mass transport through liquid state from the small to large particles.

The microstructure transformation of the fine grain Fe-Cu pseudoalloys have three stages: 1) sintering of the iron particles and the formation of the large angle boundaries in them; the growth of the iron grains and the solid state solution formation by diffusion; 2) the penetration of the liquid phase along the iron boundaries and the their desintegration; 3) repeated grain growth and the matrix structure formation, the matrix is the copper phase.

The experiments indicated that the densification of the samples was finishing during the first 10 minutes of liquid state sintering. The formation of solid iron-copper solutions were finished at the this period.

The resistivity and hardness data confirm it, - the HB and ρ values were



When during the last stage the iron grain sizes increase from 3.5 to 30-40 μm the pseudoalloys strength decreases more than by two times from 920 to 370 MPa. These data were agreed with the Stro equation.

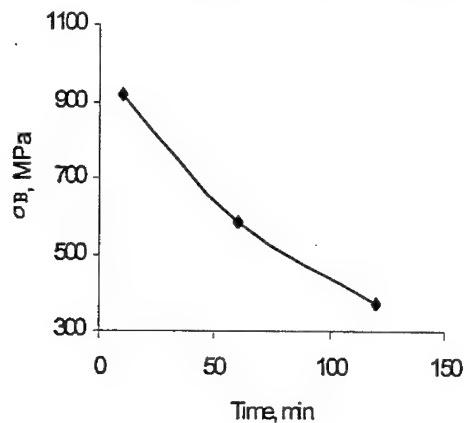


Fig 2. The strength of pseudoalloys as a function of the annealing time at 1130°C , hydrogen, the sample porosity is 1,5-2,3%

The strength of the pseudoalloys with the equal density which were sintered in solid state depend on the annealing temperature, because the iron-copper diffusion and the solid solution formation depend on the annealing temperature too. The samples which were obtained at 1000°C have the higher strength,

($\sigma_s=840$ MPa and $\delta=2,1\%$). The properties of the pseudoalloys may be improved by heat treatment, - the quenching from 1000°C and the tempering at 450°C (see table)

T, $^\circ\text{C}$	Porosity, %	ρ_0 , $\mu\text{kOm.cm}$	HB, MPa	σ_s , MPa	σ_d , MPa	δ , %
250	3,6	17,2	2190	824	738	2,0
450	4,3	9,5	2240	1020	953	2,5

When the tempering temperature was lowered until 250°C the rate of hardening had been so slow that the high strength of pseudoalloys were not obtained.



**THE FEATURES OF PROCESSES OF THE FORMATION OF A
NONEQUILIBRIUM STRUCTURAL STATE IN THE SURFACE LAYER
OF THE STEEL Kh18N10T AFTER IRRADIATION OF A PULSING
NANOSECOND LASER**

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It is known that the influence of high-energy pulses with a nanosecond pulse duration including nanosecond pulse laser can stimulate phase and structural transmutations in alloys which don't take place at traditional methods of heat treatment. The influence of a laser with the pulse duration about 20 ns is caused both by thermal action and by a shock wave (at the front of which the pressure about 1 GPa arises at a pulse energy of 1 J), capable to carry away the point defects.

The X-ray patterns (Fe K_α-radiation) obtained from the surface of the steel Kh18N10T after irradiation by a pulse nanosecond laser are characterised by splitting of all X-ray diffraction lines of γ-Fe. Along with lines, the angle position of which practically coincides with the position characteristic for the initial state (γ), the occurrence of the second very broad lines is observed at smaller reflection angles (γ_1), corresponding to the lattice constant of a FCC-solid solution, increased on average by 1,5 % against the base lattice parameter (fig.1).

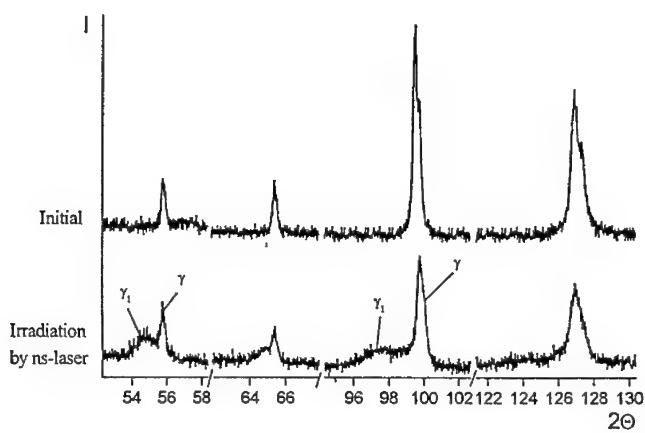


Fig. 1 X-ray diffraction patterns from the surface of specimens from the steel Kh18N10T



The radiation of the pulse nanosecond laser caused a strengthening of the surface layer in depth 10 μm that order of magnitude coincides with the depth of the existence of the two-solution nonequilibrium state. Under the influence of the irradiation by this laser a cellular dislocation structure with the cell size of 0,3 - 0,5 μm is formed in the surface layer (fig.2). The data of X-ray spectral analysis show the absence of the difference of concentration of the base alloying elements chromium, nickel and titanium in cell boundaries region and volume. The appearance of the FCC - solid solution with increased lattice parameter is imagined to be connected with the increase of the concentration of interstitial atoms formed in a great amount while shock wave action. In this process the entrainment of interstitial atoms and their non-diffusion movement in a crystal lattice are possible. It is supposed that interstitial atoms group in cell boundaries formed by dislocations. The large concentration of interstitial atoms on dislocations also inhibits the formation narrow subboundaries. Such mechanism of the formation of the new nonequilibrium γ_1 -phase is confirmed by calculations of volume fractions of dislocation subboundaries and cells themselves: the ratio of the volume of cell body to the volume of boundary regions is approximately equal to the ratio of integral intensities of X-ray diffraction lines corresponding to the reflection γ and γ_1 :

$$\frac{V_{\text{volume}}}{V_{\text{boundary}}} \approx \frac{I_\gamma}{I_{\gamma_1}}.$$

The formed cellular dislocation structure causes strengthening of the surface layer of the austenitic steel Kh18N10T. The formation of complexes "impurity - interstitial atom", of which is possible in concentrated solid solutions, shall facilitate the keeping of interstitial atoms in new position after passing of the front of a shock wave [1].

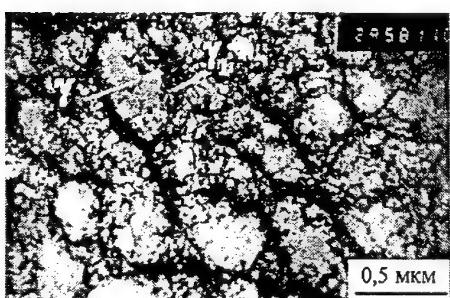


Fig.2 The structure of the surface layer of steel Kh18N10T after irradiation by a nanosecond laser

1. Robrock K.G. Interaction of own interstitial atoms with additional atoms in materials // Phase transformation while irradiation, ed. by F.W.Nolfe.- Cheliabinsk, Metallurgia, 1989.- P. 94-118.



**INVESTIGATION OF MECHANICAL PROPERTIES
OF Al-Pd-Mn QUASICRYSTALLINE SINGLE CRYSTAL
BY INDENTATION TECHNIQUE**

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Quasicrystals are very brittle in macromechanical testing to a temperature about $0,85T_m$ (Al-Pd-Mn - to 953 K), above this temperature they deform with a strong softening. At $T < 0,85T_m$ their mechanical properties cannot be determined by standard testing methods. In the report there are presented the results of works for studying mechanical properties of icosahedral Al-Pd-Mn single crystal with the help of the complex of microindentation techniques which permit to perform a plastic deformation of brittle materials and to determine their mechanical properties [1].

An $Al_{71}Pd_{21.5}Mn_{7.5}$ single quasicrystal was produced by Bridgeman technique. The crystals for investigation were cut by electroerosion perpendicularly to 2-fold, 3-fold and 5-fold directions. The following characteristics were obtained:

- deformation curves at room temperature, which are analogous to stress-strain curves obtained in standard mechanical testing, for crystals of three orientations (Fig. 1). In the course of deformation of AlPdMn single quasicrystal a weak softening is observed, which is more visible in specimens with 2-fold and 3-fold symmetry axes;

- the exponent n in the power dependence of the deforming load P on the print diagonal d at room temperature that is equal to 1.95;

- curves of the temperature dependence of microhardness for the specimen with 5-fold symmetry axis with various loads (Fig. 2). In the temperature interval of 78 – 293 K a sharp drop of hardness is observed, then a more flat section at 293 – 673 K and again a sharp lowering to a temperature of 823 K take place. It is revealed that at a smaller load in the temperature range of 293 – 573 K the temperature dependence of hardness is stronger;

- loading and off-loading curves for the specimen with 5-fold symmetry axis while nanoindentation (Fig. 3). In the loading curve there are observed small steps similar to ones revealed earlier by the authors in an Al-Cu-Fe quasicrystal and absent in metals.

In the work the deformation mechanisms are discussed. The obtained results together with the availability of several layers of extruded material at print edges (Fig. 4) give grounds to suppose the occurrence while microindentation of a phase transition to a more ductile phase.



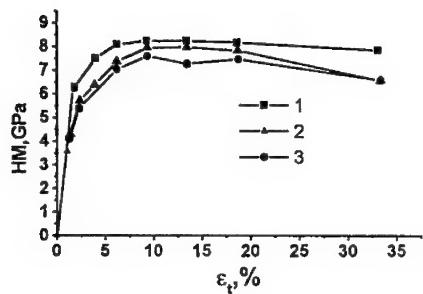


Fig. 1

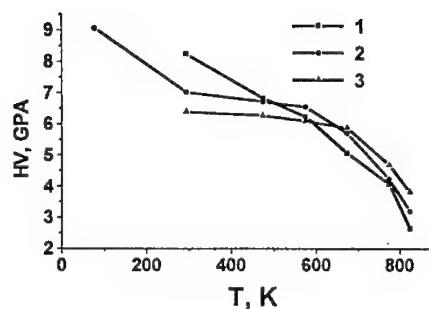


Fig. 2

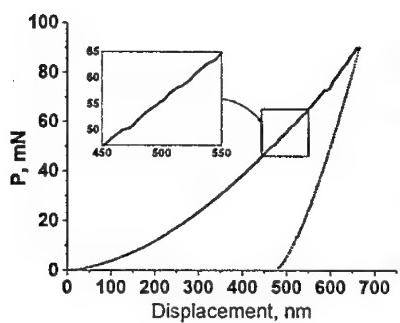


Fig. 3



Fig. 4

Fig. 1 Deformation curves for specimens with various symmetry axes: 5-fold - 1, 3-fold - 2, and 2-fold - 3.

Fig. 2 Temperature dependence of microhardness for the specimen of 5-fold symmetry axis with the loads: 1.15 N - 1, 2.35 N - 2, and 5 N - 3.

Fig. 3 Loading curves obtained while nanoindentation of the single quasicrystal of 5-fold axis.

Fig. 4 Photo of an indentation print in secondary electrons: the indentation temperature is 473 K, the load is 2.34 N.

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**STRUCTURE AND MECHANICAL PROPERTIES
OF RAPIDLY SOLIDIFIED MATERIALS THAT CONTENT BORON**

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Structure and microhardness of rapidly solidified powders of Fe₇₅B₂₅ alloy and high-speed steels on the base of the steel of R6M5 mark alloyed additionally by boron in the amount of 1,0 and 2,0 mass. % as well as of the 10R6M5 steel that contained no boron were investigated. Powders were obtained by the method of quenching from the liquid state with the help of the atomization of the melt on a water-cooled screen in the inert gas atmosphere. The temperature dependences of bending test mechanical properties in the temperature interval 20-675°C were studied for compact high-speed steels, produced from microcrystalline powders.

The grain size dependences versus the size of powder spherical particles for the Fe₇₅B₂₅ alloy and the steel of 10R6M5 mark crystallized in the inert gas atmosphere were obtained. These dependences may be approximated with the help of straight lines and may be described by the formula:

$$h = a + bd,$$

where h is effective grain size and d is the size of powder particles. The coefficient a characterizes the alloy tendency to amorphization and coefficient b is the rate of changing the effective grain size with the change of powder particle size.

For the Fe₇₅B₂₅ alloy that is amorphized easily, $a = 0$, $b = 0,020$, but for the steel of 10R6M5 mark $a = 0,64$ and $b = 0,013$ (under the condition that h and d are measured in micrometers).

For flaky powders of alloys containing ≥ 2,0 mass. % of boron, which were crystallized on the surface of a steel screen there were revealed extremely high values of microhardness in the contact surface that was in contact with the surface of the steel screen (under the measuring with the load of 0,5 N): 15 GPa for the steel containing 2,0 mass.% of boron, and 19 GPa for the Fe₇₅B₂₅ alloy, whereas for the steel of 10R6M5 mark the value of microhardness was near 10 GPa.

The metastable phases which are not characteristic for slow cooling were detected in spherical powders of steel on the base of R6M5 mark that contained about 2,0 mass. % of boron namely: two solid solutions on the base of FCC-iron with different crystalline lattice parameters (γ_1 with the lattice parameter of 2,70 Å and γ_2 with the lattice parameter of 2,52 Å), and the metastable carbide M₃C₂ as well. These phases have a rather simple crystalline lattice with small quantity of atoms in the crystal cell and a dense packings, while the stable carbide M₆C that is



the main carbide phase in the high-speed steels has a complex structure and sponging packing.

The formation of the metastable phases with the dense packing of atoms is explained by the influence of boron that has extremely high difference in solubilities in solid and liquid iron, because of which boron atoms accumulate close to the surface of crystals and hinder their growth. This facilitates the formation of the fine-grained structure. In these conditions the reduction of the interphase specific surface forces, that takes place, where the structures with the dense packings forme, will bring the energetical benefit due to the reduction of the solid phase surface energy [1].

The compact high-speed steel from rapidly quenched powders additionally alloyed by 1,0 mass. % of boron is characterized by extremely high strength in the whole interval of test temperature. The strength achieved 6.5 GPa at a temperature of 350°C, and 4.0 GPa at a temperature of 600°C.

High strength of the compact high-speed steel, additionally alloyed by boron, as well as high microhardness of flaky powders are probably connected with high interaction energy of boron atoms with the structure defects: dislocations, grain boundaries, interphase boundaries.

The estimations with the help of Cottrell formula [2]:

$$E = GbR_0^3 \varepsilon \frac{\sin \theta}{r}$$

shows that the interaction energy for boron atoms with an edge dislocation exceeds the interaction energy for atoms of carbon about nine times, i.e.

$$E_B / E_C \approx 9.$$

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UTILITY OF STRUCTURAL STATE DIAGRAMS FOR THERMOMECHANICAL TREATMENT OF METALS

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The solution of a problem of optimization of regimes thermomechanical treatment of metal materials in many respects is stipulated by installation of regularities of dislocation structure change and features of its behaviour during a plastic deformation depending on the different factors - composition, initial structural state, conditions of a deforming etc. Therefore results of mechanical tests, which are accompanied in study of these structural changes represent special value. The most convenient shape of representation of results of share mechanical characteristics and structure examinations of metal or alloy is building its structural state diagram. For the first time structural state diagrams of metals (molybdenum, chromium, vanadium) with the indicating of areas of existence of dislocation structures, sequentially appeared during a plastic deformation, represented in work [В.И.Трефилов, Ю.В.Мильман, С.А.Фирстов. Физические основы прочности тугоплавких металлов. Киев. Наукова думка. 1975].

The key opportunity of building of structural state diagrams of metals is grounded on existence of two primary factors.

At first, it is possible to view presence at a plastic deformation of step-by-step changing of one dominant type of dislocation structure by other as a result of sharp rebuilding of dislocation structure in a narrow interval of values of strain and stress, which it is regarded as critical (e_{cr} , S_{cr}). Secondly, existence of a method of definition e_{cr} and S_{cr} , in which bottom is laid in the rebuilding of stress-strain curves in coordinates $S-e^n$ (n - index of strain hardening) [В.И.Трефилов, В.Ф.Моисеев, Э.П.Печковский и др. Деформационное упрочнение и разрушение поликристаллических металлов. Киев. Наукова думка. 1989].

As a result for concrete metal or alloy it is obviously possible for sure and rather promptly to construct structural state diagrams, which reflect temperature-deformation (T-e) and temperature-power (T-S) of boundaries and areas of existence of limiting structural states in a broad interval of temperatures.

Such diagrams some views can be constructed.

ε - T - diagram DT (deformation - temperature), is restricted to area of a uniform deformation; the temperature interval (ΔT) unrestricted - from temperature of brittle failure of metal or alloy up to temperature of formation new grain structure.

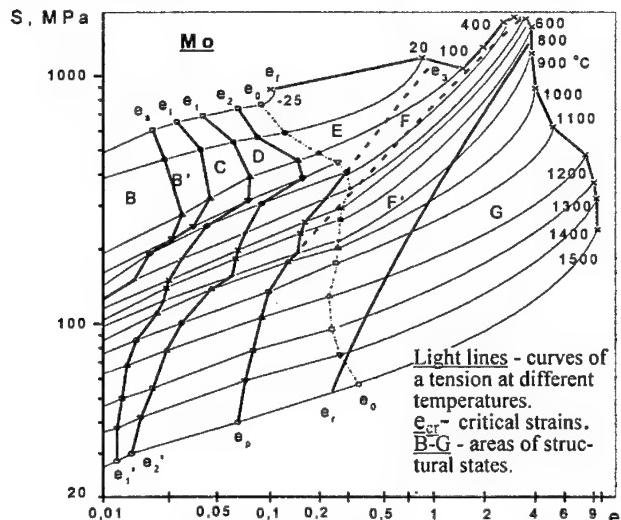
$\lg e$ - T - diagram TDT (true deformation - temperature), interval of strain - from a level relevant to a stress of a yield strength up to deformation of fracture; ΔT - unrestricted.



$\lg S - T$ - diagram TST (true stress - temperature), interval of stresses - from a stress relevant to a start of a plastic deformation ($e \approx 10^{-5}$) up to a breaking point; ΔT - unrestricted.

$\lg S - \lg e^n - T$ - diagram TSST (true stress and strain - temperature), interval of values of a stress and strain - from a level of a yield strength up to fracture; ΔT - unrestricted.

The representation of results of mechanical tests of concrete polycrystalline metal or alloy as such diagrams supplemented one another to the fully reflects temperature-deformation-power dynamic of its structural and mechanical behaviour. Simultaneously diagrams allow to identify mechanisms of a plastic deformation from its initial stages down to fracture of metal in a practically unbounded interval of temperatures in data conditions and to forecast their changes in other conditions. Thus the diagram TSST (Fig.) combines in itself two parameters of metal treatment (temperature and degree of deformation) with two its major characteristic - structure and stress, incipient in metal, i. e. in fact it is assemblage all reduced



views of structural state diagrams. Such diagrams, as it is visible, allow directly to determine concrete regimes of treatment, which can ensure the given forms or defined values of the mechanical characteristics of metal, and at the same time - to install mechanisms of a plastic deformation in concrete intervals of temperature, strain, stress and structural state.



THE STAGES OF STRAIN HARDENING OF COPPER
UNDER MULTIPLE SHOCK LOADING

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In order to describe the strain hardening curves of f.c.c.-polycrystals under tension after prior deformation the empirical expression: $\sigma = \sigma_0 + K\varepsilon^{1/2}$ (1) is commonly used.

In paper [1] the strain hardening curve of molybdenum alloy (b.c.c.) treated in coordinates true stress (σ_t) versus square root of true strain ($\varepsilon_t^{1/2}$) was compare to results of transmission electron microscopy investigation of dislocation structure on each of curve part. It was shown that certain dislocation structure types are corresponding to each of parts of strain hardening curve.

In contrast to [1], where the regularities of strain hardening were investigated under constancy of strain rate, in our paper under multiple shock loading the loading and unloading occur alternately (i.e. two processes occur – hardening and relaxation). That's why here the treatment of experimental curves (dependencies of deformation on loading time) was executed in coordinates true strain (ε_t) versus square root of time ($t^{1/2}$). I.e. the total strain of specimen must be proportional to \sqrt{t} , and as one can see in [2] experimental curves have a parabolic nature.

Such dependence ($\varepsilon = A\sqrt{t}$) we also can obtain from Orowan's equation: $\varepsilon = ab\rho S$ (2), where a is coefficient approximately equal to 1, b is Burgers vector. Let's suppose that middle dislocation path (S) relates with dislocation density (ρ) as: $S \approx l/\sqrt{\rho}$, and dislocation density changes with time by following way: $\rho = nt$, where n is the number of dislocations that adds during deformation to volume unit during time unit. Taking into account those suppositions we obtain the following: $\varepsilon = ab \cdot nt \cdot (1/\sqrt{t}) = ab\sqrt{nt} = A\sqrt{t}$ (3).

As one can see from fig.1d the strain hardening dependence treated to dependence $\varepsilon_t = f(t^{1/2})$ is the broken line with three straightforward parts.

The detailed study of dislocation structure evolution of metals including copper after multiple shock loading with several frequencies was carried out in [3], here the general regularities of dislocation structure changing that formed under shock treatment were determined.

In our paper the transmission electron microscopy investigations of dislocation structure of copper specimens that correspond to each of three straightforward parts of dependence $\varepsilon_t = f(t^{1/2})$ show the followings:



- when $\varepsilon_t = 0,03$ (the first part) we observe the chaotic arrangement of dislocation - "wood" and increasing of dislocation density up to 10^{10} cm^{-2} while the strain increase up to $\varepsilon_t = 0,08$ (fig.1a);
- when $\varepsilon_t > 0,1$ (second part) we observe the clew dislocation structure (fig.1b);
- when $0,2 < \varepsilon_t < 0,3$ (third part) – the cell dislocation structure is formed (fig.1c).

Thus, in our paper was determined that each of three straightforward parts of dependence $\varepsilon_t = f(t^{1/2})$ for copper corresponds with certain dislocation structure type: I – "wood"; II – "clews"; III – "cells" caused the strain hardening rate and the dependence $\varepsilon_t = f(t^{1/2})$, as those one proposed in [1], permits to reveal the stages of strain hardening.

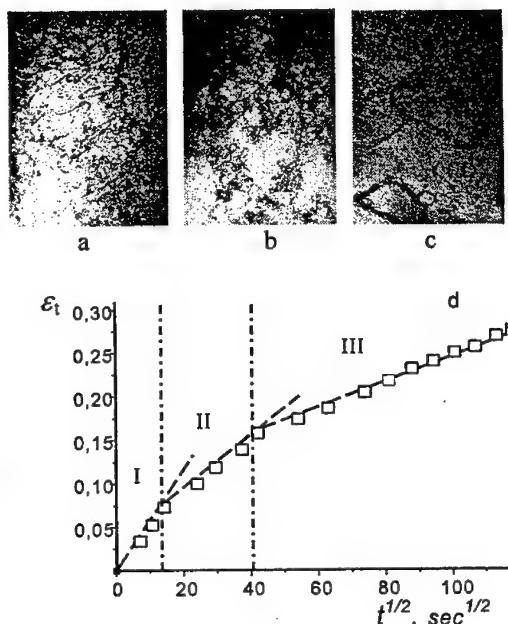


Fig.1. The dislocation structure evolution o copper during low-frequency shock treatmen (0,8 Hz) (x19000): a) $\varepsilon_t=0,03$; b) $\varepsilon_t=0,1$; b) $\varepsilon_t=0,25$; r) strain hardening curve i coordinates $\varepsilon_t - t^{1/2}$.

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ELECTRO- AND ACOUSTOSTIMULATION OF THE PLASTIC DEFORMATION BY TWINNING

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The electrostimulation of plastic deformation of metals and alloys at simultaneous influence of mechanical load and high-density electric current impulses is reliably established. Despite of numerous of publications dedicated of this problem the physical nature of electroplastic effect till now is not clarified.

It is known, in both cases at slip [1], as well as twinning [2] the main operating of short-lived current pulses has non-polar nature. It has been obtained that displacements and multiply of perfect dislocations and additional growth of twins do not depend on a direction of a current pulse, which goes through the crystal.

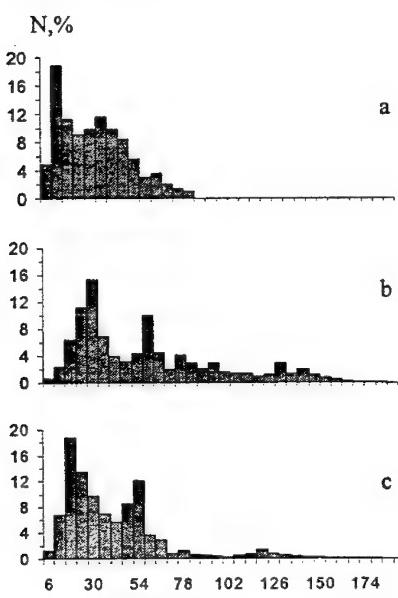
Twinning in metals is always accompanied by slip, which causes that the generation of a mechanical twin and its subsequent grows under the action of a raising stress has probabilistic nature. Interaction of slip and twinning provides the conditions, under which lengths and thickness of wedge-shaped twins created by indentation load are change by random way. This phenomenon is quantitative describe by means of the statistical twins sizes distributions and density twinning dislocations at twin boundaries distributions. The statistical characteristics of the twins lengths distributions are very sensitive to conditions of deforming. Their comparative analysis allows to establish elementary dislocation processes determining development of an ensemble twinning dislocations which are created a twin layer at different loading [3].

The study of twins distribution for several hundreds impressions in experiments on indenting of a single crystal of bismuth at room temperature has shown, that at simultaneous operating of steady load and electric current impulses it is impossible to select some of three crystallographically equivalent pairs of planes and directions twinning, in which one twins would receive primary development under the influence of an impulse. Has occurred a supposition, that by operating of current pulses is created under a sample of short-lived dynamic stresses.

For checking up this hypothesis three series of the experiments were realized. At all experiments imposed load produced the identical increase of effective internal stresses in the neighbourhoods indenter at dimpling of the cleavage plane of bismuth single crystal. Evidence of that is identical increase of the diagonal of the indenter impressions. In the former case through the twinning crystal located under steady load 0,1 H the current impulse with density $i=500 \text{ A/mm}^2$ and duration 10^{-5} s was passed. The increase of diagonal of the indentation corresponded to operating of steady load 0,2 H, in this connection a series of twins at the same



static loading were studied. In the third case through the sample under power 0,1 H the acoustic impulse of the same intensity and duration, as electric impulse was passed. The histograms of the twins lengths distributions under influence of steady load (a), steady load both current impulse (b) and steady load and acoustic impulse (c) are illustrated in Fig. 1. The first distribution is constructed by results of 597 measurements, second and third ~ 900.



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INFLUENCE OF COLD ROLLING ON MECHANICAL PROPERTIES AND
DESTRUCTION OF STEEL 45XH2MФA AFTER STRENGTHENING
ELECTRIC-THERMAL TREATMENT.

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The most perspective ways of high strength steel goods obtaining relates now with creation of technological scheme used high heating rates. Combination of cold rolling with several variants of high rate electric thermal treatment into one technological cycle discover the outlooks of increasing both the technology and economical production figures and improving of characteristics complex of goods due to display the effect of inheritance strain structure elements that contributing to additional hardening after final treatment. It is important to choose the optimal strain during single reduction and also to determine the total strain that not causes the danger structure defects appearance.

The aim of present work was the determination of limit strain of 45XH2MФA steel under cold rolling and also the possibilities of latter to properties improving after thermal-hardening by scheme fig.1 by using electric heating.

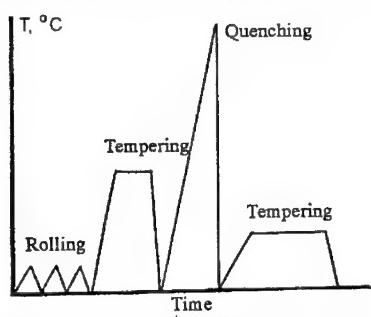


Fig.1. The technological scheme of goods production.

The longitudinal rolling was carried out with different values of single reduction: 2...3, 5...7 and 9...11% up to total strain 41, 53 and 63%, respectively. The studies of internal structure and destruction surface structure were carried out by optical, scanning and transmission electron microscopy. Mechanical properties of steel in thermal hardened state after several variants of treatment are shown on fig.2.

By above described investigations was shown the followings:

1. Plastic deformation by cold rolling in subsequent thermal hardening not create studied regions of reduction and suppositions that contributing to destruction.
2. The destruction of thermal hardened specimens is reasoned by mechanical origin surface defects in 20% cases, by pores (fig.3a,b) in 40% cases and by intercristallite bursting that occurs under testing in room temperature region (fig.3c,d) in 33% cases. In 7% cases evidently marked centre of destruction are not revealed. In last case the mechanical properties mount to maximum values whereas in all other cases it's lower and sometimes it's not mount the yield stress.

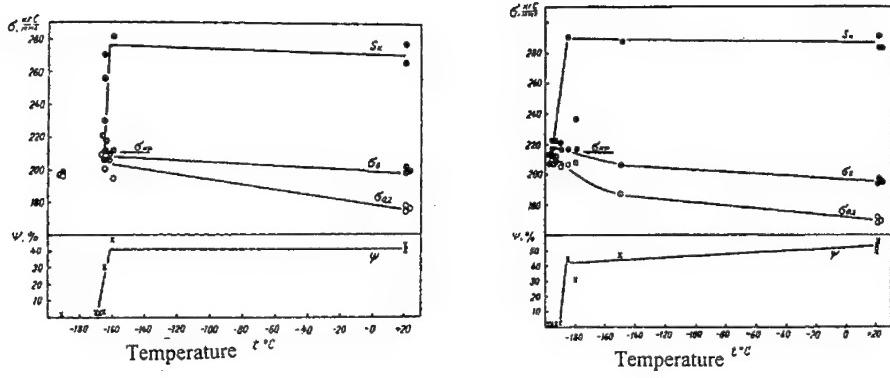


Fig.2. The results of mechanical testing of thermal hardened 45XH2MFA steel without rolling (a) and with rolling (b); $e_i=5\ldots 7\%$, $\varepsilon_i=53\%$.

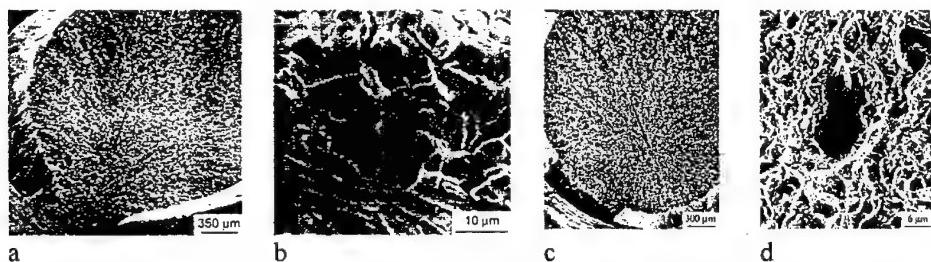


Fig.3. The fractography patterns with centred nature of destruction surface of thermal hardened 45XH2MFA steel without rolling (a,b) and with rolling (c,d). The centres of destruction are shown by arrows, b,d – the same places respectively (a,c).

3. The prior cold plastic deformation in all studied cases conducts to decreasing the destruction temperature (when ψ sharply decreases) on 20 °C. And the yield stress increases on 5...7 kg/mm² in comparison with unstained state.

4. The stable realisation of maximum opportunities of 45XH2MFA steel can't be achieved without elimination of above mentioned defects that lead to strikingly expressed centred destruction.



RHEOLOGICAL MODEL OF A RESIDUAL TWIN LAYER IN METAL

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Mechanical twinning is totally responsible for the strength and plasticity of many metals and their alloys widely employed in technology. The propagation of a twin layer can stimulate crack initiation and cause the destruction of a material. And vice versa, as a result of rapid growth a twin is able to reduce concentration of internal stresses, promoting their relaxation. It is precisely impossible to forecast a development manner of the deformation twin at different modes of loading, because of strong influencing on this process of different factors.

While analysing plastic deformation of calcite and sodium nitrate crystals [1] Garber concluded that under a load the single residual twin layer behaves like a macroscopic sample exposed to deformation. In a consequence this idea was reliably confirmed and developed in the investigations of other scientists.

In the experiments on studying of twinning metals at different modes of loading are detect: elastic retardation, hysteresis, flowability, creep, hardening and softening, relaxation phenomena, Baushinger effect [2]. Consequently, at loading and unloading of twinning metallic crystal a residual twin behaves as a elastically-viscous-plastic body and has all fundamental rheological characteristics: elasticity, plasticity, viscosity and strength.

On the basis of a macrorheology [3] the mechanical behaviour of the twin layer into the crystal under power it is possible visually to present by means of rheological model of a standard solid, which is shown in Fig. 1. Here springs are present elastic, and cylinder piston – plastic properties of a material. The rheological parameters of this model have the following physical sense: G_1 is some elastic module inherent of the twin boundaries, G_2 – elastic module of a sample matrix, η_1 – coefficient of viscosity of the twin boundaries motion.

By virtue of rheological model the main equation of stress-strain state of the twin layer may be obtained. For the model of a standard solid it has a form similar to the equation of the standard linear body:

$$\sigma + t_\varepsilon \frac{d\sigma}{dt} = G(\varepsilon + t_\sigma \frac{d\varepsilon}{dt}), \quad (1)$$

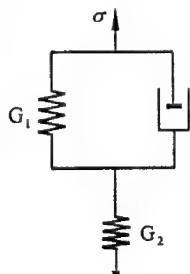


Fig. 1



where σ – applied stress, ε – strain, which is equal for the twin deformation to

$\varepsilon = \frac{\Delta d}{d_0}$ (d_0 – initial the twin thickness, Δd – displacement of single twin boundaries after loading),

$$\tau_\varepsilon = \frac{\eta_1}{G_1 + G_2}, \quad t_\sigma = \frac{\eta_1}{G_1}, \quad G^{-1} = G_1^{-1} + G_2^{-1}. \quad (2)$$

This parameters (2) have the following meaning: G is the relaxed elastic module of that model (fig. 1), τ_ε – relaxation time of stress at permanent deformation and τ_σ – relaxation time of deformation at constant stress.

Solving of the rheological equation of state (1) for various modes and stages of loading one may to compute the quantitative characteristics of Baushinger effect, local damping decrement and to make the numerical estimation of the value of the effect loss of hardening on the wedge-shaped twin boundaries under pulse loading. Calculations (1) by different functions $\sigma(t)$ are enough difficult.

However, if the experimentally fixed modes of pulse or alternating-sign loading to approximate by more simple periodic laws, it will allow considerably to simplify the solution of an equation for steady-state process and to simplify his physical treatments.

In particular, for the quantitative measure of Baushinger effect the integrating of an equation (1) gives:

$$\beta = 2\sqrt{2} \frac{G_2}{\sqrt{\eta_1 \sigma}} \varepsilon_0^{3/2}$$

where β – displacement of the twin boundaries at alternating-sign loading, ε_0 – twin deformation by stress σ_0 , by which loading is interrupted.

Numerical estimations of values parameters G_1 , G_2 , η_1 and β give a good agreement with experimental data. The obtained results show that this phenomenological descriptions for the behaviour of the twin boundaries under alternating-sign loading is adequate, that allows rheological modelling to be applied for the investigation of mechanical properties of a residual twin layer.

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SIMULTANEOUS inCREASING of STRENGTH and PLASTIC CHARACTERISTICS of CAST TITANIUM-ALUMINIUM ALLOYS BY THE FORMATION of CAST STRUCTURE of α -PHASE

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It is established, that the characteristics both strength and plasticity of cast titanium-aluminium alloys can be increased not only by high homogeneity and uniformity at reduction casting defects of ingots, but also by modeling of cast structure of α -phase.

On an example of a cast alloy VT-5C, received by the method electronic beam skulled smelting (EBSS) with electromagnetic mixing melt and high speed of crystallization, it is shown of principle possibility of change the "classical" structure cast VT-5C alloy, received by a method of vacuum-arc smelting (VAS). The microstructure of an VT-5C alloy, melted without mixing in conditions VAS, is considerably distinguish in the case of imposing of electromagnetic fields at EBSS. As a result of intensive mixing an alignment of concentration alloyed elements all over the volume is happened (in given case it is an aluminium), and the rather large speed of cooling at pouring in the graphite or copper forms prevents the diffusion mobility of atoms. If in the first case the origin α -phase plates occupied boundaries of grains and during the further cooling these plates sprout through all volume of a grain in one direction, in second -one occurs simultaneously in all volume of a grain. Just such disposition of α -phase plates also results to simultaneous increasing of strength and plasticity.

The mechanical properties of a cast alloy received by a method EBSS, considerably exceed ones of an VAS alloy, (the shock viscosity is $0,5\text{-}1\text{Mn.m}^{-1}$ at admitted for this alloy - $0,3\text{Mn.m}^{-1}$).

It has been studied the casting, received by a method electronic beam smelting (EBSS) at the following technological modes - vacuum $10^{-2}\text{-}10^{-3}\text{MPa}$, having overheated under melt up to $2073\text{-}2173\text{K}$, the electromagnetic mixing of the melt under the current frequency of the power supply $-2,4\text{kHz}$. The thermodynamic parameters of the fusion have allowed to ensure more deep refining titanium alloys from gas impurity, than ones at VAS. In alloys VT-5C EBSS hydrogen content is within $1\text{-}3 \cdot 10^{-3}\%$, that in 10-100 once below in contrast with the metal, received by the method VAS. Thus it was possible to reduce a contents of impurities O and N to a level 0,05 and 0,01 % accordingly.

The developed technological parameters melting and casting have allowed to receive not only high purity from impurity inculcation (total contents $< 0,1\%$), high homogeneity of chemical composition titanium alloys.



FRACTURE MICROMECHANISMS AND FRACTURE TOUGHNESS OF TI-SI-X AS-CAST COMPOSITES

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Ti-Si-X alloys alloyed additionally with Al and Zr are highly strong and heat resistant materials those could be as a perspective base to develop novel composites of different application.

Fracture mechanisms were studied with scanning electron microscopy of samples tested on fracture toughness in tempearture range RT-800°C.

As the first feature of brittle fracture of the cast Ti-Si-X alloys is an independence of a place of its origin on a notch. The cracks start from both the notch and its opposite side of sample. Fractographical analysis shows that this effect is promoted by not only structural defects of casting but large particles of α -Ti-phase precracked in a course of loading. By the second one is the important thing that even large enough silicide particles being microcracked do not delaminate themselves from α -Ti-matrix. This points out the high adhesive strength of particles with matrix that realizes in good high temperature strength of the Ti-Si-X alloys.

Fracture mechanism in all the temperature interval studied is cleavage with some portion of ductile pores (voids) coalescence.

Data obtained show also that fracture toughness of the Ti-Si-X alloys determines basically by silicon concentration in them: the lower silicon, the better fracture toughness. As the second factor influencing on fracture toughness is size and morphology of hard particles of the second phase. It is known that in dependence on toughness of matrix, in a course of pre-deformation particles may be as both the sources of crack origin or obstacles for cleavage crack spreading.

As follows from fractographical analysis data, the rough silicide particles of eutectic colonies fail by cleavage at all temperatures that may explain the low figures of fracture toughness. In alloys where matrix has some plasticity and contains fine particles some features like small separate dimples may be found in fracture surface that evidences on failure with pores coalescence. At temperatures above 500°C, ductile features occupy the large part of surface that results in higher fracture toughness.



INFLUENCE OF MICROALLOING ON CHARACTERISTIC
STRENGTH AND PLASTICITIES OF A LOW-ALLOYED CHROMIUM

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Influence of little additives (0,5 wt.%) of zirconium and renium on structure and properties of sheet material low-alloyed chromium was studied to rise of strength and decrease inclination of alloys to dynamic deformation ageing at the same time increasing technological plasticity.

Successes of studing of dynamic deformation ageing and recrystallization embrittlement of chromium and its alloys allowed us to create the new low-alloyed alloys VKh-2K-2, VKhM, VKhMR with raised low-temperature plasticity. Complex alloying of La, Ta, Zr, Re provided high mechanical properties to alloys of this types due to combination of deep refining of matrix and its to solid solution hardening.

Research of structure of VKhM alloy showed that additional alloying by low amount of zirconium (0,1%) led to decreasian of quantity of unconned carbon and to formation of sirconium carbide dispersoid thus improyng properties of the alloy. A relative elongation at the temperature 1073 K achieved 25-30%.

Distrengthen of sheets of the VKh-2K-2 alloy begins to display after heat treatment at the temperature 1273 K during one hour. However sharp distrengthen of sheets it is observed only after annealing at the temperature 1373 K. A yield of strength is lowered from 680-750 MPà to 280-300 MPà. Necessary to mark that the beginning of distrengthen of sheets does not coincide with the beginning of recrystallization.

It is estaeablished that under structure research of annealing samples at the temperature 1223 K, typical structure for deformed state -fibrous structure and fragmentation of polycristalls by cells is remained, the boundaries of cells become more thin. Research of annealing samples showed that recrystallization of the alloy takes place after annealing at the temperature 1373 K during one hour. A grain dimension preserves almost without changes up to 1573 K. Sheet material has the most plasticity after annealing at the temperature 1373 K. Microhardness herewith falls down to 1500 MPà. Prolonged heat (1373 K, during 10 hours) allows to raise lowtemperature plasticity of the alloy in recrystallization state. Herewith a relative elongation reaches up to 30-35% due to disintegration processes flowing with formation of disporsion inclusions.

An optimum combination of strength and low-temperature plasticity of sheet material is reached by the following heat treatment: polygonization annealing



+ recrystallization. Finish heat treatment provides deep disintegration of solid solution and even apportionment of particles of admixture phases of high active alloying elements in metal structure. Duration of heat treatment of ready sheets is determined by the speed of disintegration process and grain dimension and is equals 5-10 hours.



INFLUENCE of NITROGEN ON the MECHANICAL CHARACTERISTICS Cr AT the INCREASED TEMPERATURES

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If the most dangerous impurity for Cr, as well as for the majority of metals, is the oxygen, which can raise temperature of cold brittleness up to 500°C [1], the action of carbon and nitrogen is displayed to a lesser degree. So the introduction in an alloy of nitrogen in quantity up to 0,3-0,4 wt% raises temperature cold brittleness only up to 200-250°C [2]. The danger influence of nitrogen consists of high temperatures of forming nitrides after actively penetration through crack and pores of formed oxygen film. At high temperatures the nitrogen is actively diffused in a matrix, concentrated at the beginning mainly on boundaries of grains as thin long needles.

The research of initial stages of interaction between nitrogen and chromium has shown, that the homogeneous second phase begins to fall out us occurrence prior allocation of thin disks by thickness in some internuclear distances on planes Cuba {100} at temperatures 600-650°C. However stability nitrides prior allocation is not so high and at temperature 900°C and above occurs coagulation of collection of atoms of nitrogen by massive allocation with significant fields of microstress near them [2]. After forming of massive nitrides high-temperature strength of an alloy falls, but at a temperature interval 600-900°C, and sometimes up to 1050°C (temperature of aging depends on a choice alloyed of elements), the strength of an alloy raises and this increasing can be very significant (fig.1).

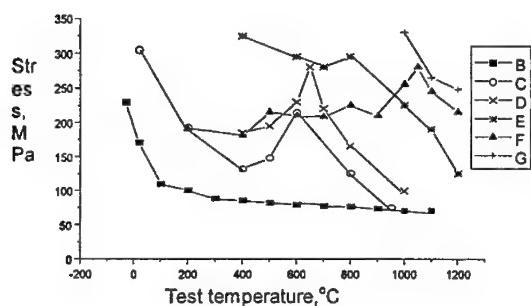
On fig.1 the temperature dependence of the characteristics of strength of cast La contented metal, double alloys with various quantity of nitrogen and multicomponent alloys are submitted. From the given data it is visible, that the introduction of nitrogen with various alloyed components can create rather wide area, which have characteristic of strength essentially higher, than its for pure chromium (alloy Cr-0,5La).

On fig.2 the changes of a limit of yield strength and fracture strength, and also relative lengthening of cast low alloyed alloys under annealing a stretching condition against of amount of the entered nitrogen and tested by at temperature 900°C are given. With increasing of amount of the nitrogen, entered into an alloy, on a graph in half-logarithmic coordinates, the increasing of the characteristics of mechanical properties in the investigated quantitative intervals carries linear character, that can testify about their exponential dependence.

So, the introduction of nitrogen though it is raised temperature of cold brittleness up to 200-250°C, in a temperature interval 600-1000°C it can essential



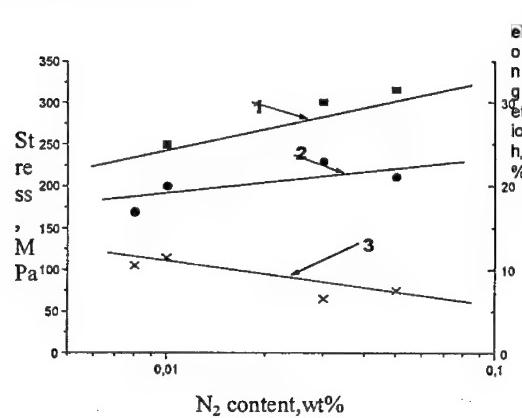
increase high-temperature strength of an alloy, and additional alloying it is possible essentially to lower temperature of cold brittleness.



B – Cr-0,5La; C – Cr-0,02N₂; D – Cr-0,3N₂; E – Cr-0,3N₂ (deform.); F – BX-2K+0,2N₂; G – Cr-AlN (evt.)

Fig. 1 A yield strength of cast Cr alloys with the different contents of nitrogen at various temperatures of test.

However it is necessary to notice, that the formation prior allocation chromium nitrides causes of processes of dynamic deformation of aging [2], that can also result to essential decreasing of the characteristics of plasticity in this area of temperatures.



1-fracture strength, σ_B ;
2-yield strength, σ_{02} ;
3-elongation, %

Fig. 2. Dependence of mechanical properties of Cr against the contents of nitrogen in an alloy (temperature of test 900°C).

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MAKING OF ANTIFRICTION COMPOSITES WITH METALLIC
MATRIX AND DYNAMIC SYNTHESIS DIAMONDS AS REINFORSING
ADMIXTURE UNDER ACTION ELECTRODISCHARGE SINTERING

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Antifriction composites (here and further on text *AC*) may be using for making of bearings or sliding supports. Their main guaranteeing destination - low sliding coefficient and small losses on friction at increase durability of united surfaces. The main *AC* estimation criterion there is his friction coefficient on surface of united bodies.

In generally, all materials for sliding bearings there are alloys on copper base. A copper has high heat- and electroconductivity, plasticity and corrosive steadiness. On state and properties of copper composites of different nature may to have influence the quantity of admixtures at alloy. The may to change properties of sliding surfaces. For increasing serviceability of copper composites is necessary to adding into volume of *AC* a nickel in quantity upto 12 % from mass, but no more. For more of nickel quantity is observed lowering of serviceability by reason of decreasing heat conductivity of matrix. The zinc concentration can to have influence on serviceability of *AC* also. But her influence may be only after 5 % from mass in matrix composite.

As *AC* base can to use aluminum also. For investigations of *AC* base (metallic matrix) was selected powder of high hard aluminum alloy V95 type with size of parts $\sim 100 \mu$.

For special conditions of work, when is present press out from friction surfaces of plastic and liquid smearing, need to have defense of friction surfaces from their reinforced wear and grip. For this using the hard inorganic matters as dry or hard smearing. From alls known materials in nature, best properties has a diamond. He have high heat conductivity, stoutness and lower friction coefficient. This aspect was stimulated his choice as reinforcing admixture to *AC*. But diamond is abrasive material and during of act work can be as cut instrument (i.e. can hasten friction steams). That's why for made of *AC* recipency use of high disperse dynamic synthesis diamond powders of 200-500 nm [2]. Their dimensions of parts dismiss scratching possibility of attended surfaces [3].



The electrodischarge sintering of composites is made with help equipment's STRUM-901A and ERAN 2/1 by power 40 kW each. Made the samples of two types:

- in asbestos cement press-tools with metallic electrodes-puncheons (STRUM-901A) with dimensions of samples 7×7×24 mm under action of alternate electrical current of industrial frequency (50 Hz) and with following contents of metallic matrix, from mass: 78 % Cu, 6 % Sn, 8 % Ni, 5 % Zn and 3 % Si with admixture 1 % from volume of samples dynamic synthesis diamonds.
- in graphite press-tools (ERAN 2/1) with diameter of samples 9 mm and height 8 mm on base of aluminum alloy V95 type with following contents of metallic matrix, from mass: 4.4 % Mg, 1.36 % Cu, 0.06 % Fe, 6.81 % Zn, 0.14 % Si, 0.14 % Cr, 0.01 % Ti, Al the rest and with admixture 3 % from volume of dynamic synthesis diamonds. Electrodischarge sintering is made under action of superposition direct and alternate high frequency (5000 Hz) electrical currents.

The theoretical computations and experimental measuring of speed resilient waves distribution and their comparison was obtained also.

The dynamic synthesis diamond powder as antifriction admixture can to give new possibilities for problem of receipt and application new types AC and to decrease the main parameter of AC - a friction coefficient and relative wear. This is new AC on copper base with admixtures of dynamic synthesis diamond powders what is obtained with help electrodischarge sintering have a friction coefficient 0.08, relative wear 5.6 μkm , and on aluminum base (V95 type) 0.11 and 2.1 accordingly. For comparison alloy V95 without admixtures has a friction coefficient 0.13, and relative wear 8.3 μkm .

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**INCREASE IN THE OPERATION CHARACTERISTICS
OF WORKING SURFACE IN THE INTERNAL COMBUSTION ENGINE
PARTS**

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Cast iron ingots are widely practised in the tractor and automobile constructions industry. Their using are especially effective in the condition of friction and wear and tear. This is explained peculiarities of their structure and composition. In common place increase in the operation characteristics are achieved by using high cast alloying elements, such as Ni, W, Mo. In this paper was studied the possibilities of using Cu instead of Ni in the condition of direct crystallization with the aim to increase wear resistance chilly surface "heel" in the "rocking arm" ingot.

It's well known that Cu increases strength, toughness and hardness of cast iron. Due the graphitization influence of Cu within ledeburite temperatures and carbide stabilization of Cu within eutectoid transformation temperatures, Cu avoid the formation of ledeburite cementite of sections and promotes the formation of ferrite in thick sections of ingots. Cu lessens the sensibility of ingots to thickness and gives opportunity to obtain more homogeneous system in all volume structure.

The analyses of metal structure of experimental cast iron in show the pearlite, the grain size of experimental and base cast iron are similar. The microhardness of pearlitic matrix are higher (10-15%) then microhardness of the base cast iron. More high heat current is also the reason of increase the wear resistance in 3-5 times against the base cast iron.

Formation of the striking example of white zones and grey zones in ingots, maximum chilly depth in "heel" and free from ledeburite cementite pearlitic metal matrix are the main favorable factor in the problem of increase of wear resistance of experimental cast iron.

The metallographic investigation are shown that the experimental cast iron has austenite frame structure in zone of full chilling (Fig.1). The lamellar type of ledeburite eutectic predominates over honeycomb type of ledeburite eutectic. Pinacoidal crystal structure of cementite is additional factor in the increase of wear resistance.

In fig.2 it has been shown that during irradiation of cementite with a laser pulse the evaporation of this phase proceeds layer by layer.



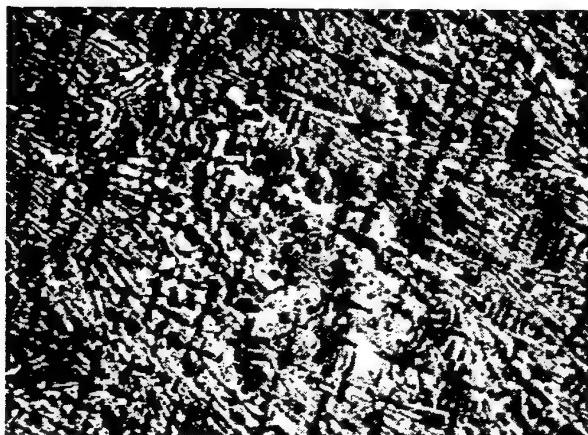


Fig.1. Direct
dendritic austenite
structure in the
chilly iron ingot
($\times 100$).



Fig.2 Stratified
structure of cementite
grain after evaporation
during irradiation with
a laser pulse ($\times 20000$)



SEARCHING FOR METHODS OF IMPROVEMENT OF STRENGTH AND FRACTURE TOUGHNESS OF WELDED JOINTS ON ADVANCED ALUMINIUM-LITHIUM ALLOYS

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The paper deals with the methods of improvement of the strength and fracture toughness of welded joints on aluminium-lithium alloys of different alloying systems (Al-Li-Mg, Al-Li-Cu, Al-Li-Cu-Mg). Alloys differ from other high-strength aluminium alloys by a combination of a high modulus of elasticity and low density. Application provides a reduction of the structure weight and promotes higher performance compared to the traditional alloys. This allows including these alloys into the class of the most promising materials for aerospace engineering. Their successful introduction into industry largely depends on the ability to produce tight welded joints of equivalent strength. A common feature of all the aluminium alloys with lithium is an enhanced proneness to coarse porosity in the welds, this being caused by a high content of gas-forming compound in the surface layers of the semi-products. Therefore, it is first of all necessary to prevent the formation of excessive porosity. Gas pores appear in heating and surface melting of aluminium-lithium alloys. These are most often concentrated in the weld lower part. It is especially pronounced in the case when the degree of the plates penetration is insufficient stirring of the molten metal in the weld pool. The second problem in welding aluminium-lithium alloys, consists in the need to limit their susceptibility to hot cracking in welding. Crack form in the deposited metal and are of an intercrystalline type. The next problem in welding of the considered alloys is the need to improve the strength and fracture toughness of the welded joints, especially in arc welding processes. The disadvantage of aluminium-lithium alloys, compared to other heat-hardenable alloys, is a higher brittleness of the metal in the fusion zone, or, in other words, a greater sensitivity to stress raisers. Tensile tests of samples with notches and a fatigue crack that with the increase of the degree of sharpness of the stress raiser, the values of the critical coefficient of stress intensity welded joints are reduced. Performed investigations allowed establishing the dependence of the joint properties on the technological factors, including heat input with different welding processes, intensity of the weld pool mixing, and application of more effective welding consumables. Obtaining better results is promoted by application of such techniques as pulsed-arc, laser and electron beam welding, modifying the weld metal structure by applying special welding wires with scandium, zirconium and other microadditives, and heat treatment of weldments. Improvement of properties is achieved not only in the weld, but also on the



boundary of the weld fusion with the base metal. The developed technological solutions open up the prospects for implementation of the advantages of aluminium-lithium alloys in fabrication of welded structures for aerospace engineering.



TREATMENT OF CUTTING PLATES OF ALLOY BASED ON TUNGSTEN CARBIDE BY PULSED MAGNETIC FIELDS

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At present pulsed magnetic fields are widely applied in various areas of a science and engineering. One of directions of application of pulsed magnetic fields in the field of engineering is change of operational properties of cutting tools by means of influence on them a sequence of pulses of a magnetic field (intensity $\sim 10^5$ A/M). It is established, that the hardness and the wear resistance of these products raise, microtensions are changed, allocation fine disperse carbide phases are observed after such influence on tools of high speed-cutting steel [1].

Alloys based on tungsten carbides are objects of rather not numerous researches in this field (as against high speed-cutting steels). It is possible to find some similar features in structure and composition by comparison of these steels and alloys based on tungsten carbides: presence of carbide phases and a ferromagnetic component of the same components. This fact probably allows the established and experimentally confirmed for high speed-cutting steels mechanism to use for the alloys (change of structure, residual tensions).

Plates of T15K6 alloy were treated by pulsed magnetic field (PMF) and passed industrial tests on Druzhkovsky machine-building plant on operations cutting and milling. Tests have shown increase of the period of stability treated plates on operation cutting – in 1,7 times, on operation milling – in 1,3 times. The X-ray analysis of samples has shown reduction of interplane distance of (111) line of a cobalt phase after PMFT.

There are various imaginations and models of influence of a pulsed magnetic field on materials.

The authors of work [2] have established, that owing to action of PMF there is a deformation caused by own fluctuations of researched object, and an increase of a field intensity it is deformation of a stretching that reveals a complex character of distribution of tension in the volume of the object.

The influence of a pulsed magnetic field on a cylindrical conductor with conductivity of a the alloy and magnetic field aperiodic impulse parameters of used in the PMFT technology is calculate in the work (in the assumption of the separated task).

The basic equation of distribution of an electromagnetic field is the equation of diffusion (neglecting currents of displacement in comparison with currents of conductivity).



Having solved the task of diffusion of a magnetic field in conducting cylinder of radius R (assuming, that the field is distributed much more slowly, than a sound), placed in longitudinal magnetic field $H_0\phi(t)$ by means of a standard method [3], we shall receive:

$$\frac{\partial^2 H}{\partial r^2} + \frac{1}{r} \frac{\partial H}{\partial r} = \frac{1}{a^2} \frac{\partial^2 H}{\partial t^2}; 0 \leq r < R, 0 < t < \infty,$$

$$H(r, 0) = 0, 0 \leq r < R,$$

$$H(R, t) = H_0\phi(t).$$

$$H(r, t) = 2 \sum_{n=1}^{+\infty} \frac{J_0\left(\frac{\mu_n r}{R}\right)}{\mu_n J_1(\mu_n)} \frac{a^2 \mu_n^2}{R^2} e^{-\frac{a^2 \mu_n^2}{R^2} t} \int_0^t e^{-\frac{a^2 \mu_n^2}{R^2} \tau} H_0\phi(\tau) d\tau$$

Where H - magnetic field, r - radius of a sample, t - time, J_0 and J_1 - Bessel functions, $a^2 = \frac{c^2}{4\pi\mu\sigma}$, c - speed of light, σ - conductivity of a material, μ_n - root of equation $J_0(\mu_n) = 0$, τ - variable of integration.

With reference to concrete conditions of technological process let us take the function describing aperiodic impulse as $\phi(t)$:

$$H(t) = H_0(e^{-at} - e^{-bt}),$$

and characteristic parameters $t_h = 200 \mu s$, $R = 1 sm$, $\sigma = \sigma_{alloy}$, accept μ is equal one.

The theoretical calculations allow to give ponderomotive forces distribution in volume of a sample as function of time.

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**INFLUENCE OF SUBSTRATE ON FORMATION OF DEFORMATION
ZONE, DURING INDENTATION OF METALLIC GLASS TAPE BY
VICKERS PYRAMID**

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The microindentation of thin metallic glass (MG) tape has special features [1]. They are caused by little thickness and high strength of MG [2]. As a result mechanical characteristics of substrate influence on results of microindentation.

The purposes of the research: (1) investigation of formation of deformation and destruction during indentation of annealed MG situated on substrates characterized by different mechanical properties and (2) investigation of the substrate role in the process of MG indentation by pyramid of Vickers, on the base of model experiments.

The material used in the present work was MG 82K3XCP. The thickness of the tape was 30 μm , the content 83,7%Co+3,7%Fe+3,2%Cr+9,4%Si (wt.%). Before testing samples were annealed in the temperature range of $T_{an}=373 - 888$ K. The character of MG deformation and endurance to cracking was investigated by indentation. Three different polymers (microhardness: № 1 \approx 151 kG/mm 2 ; № 2 \approx 16 kG/mm 2 ; № 3 < 10 kG/mm 2) were used for substrates. The thickness of substrates was 1 mm. Substrates situated on strip of metal.

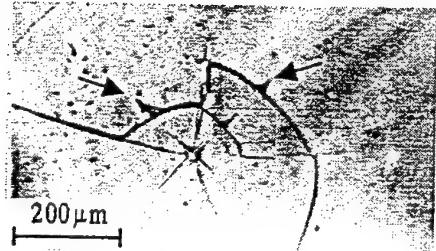
The increased copy of Vickers pyramid was made for model experiment. The materials with different mechanical properties were used for preparation of substrate. Metal foil was used as a covering.

Indentation of not annealed MG by loads 100 – 400 g is accompanied by formation of deformation zone. Experimentally found formation of similar zones after model experiments. It is established that observed morphology is caused by formation of pleats from MG around place of indentation.

The morphology of destruction is changed as a result of annealing temperature growth. The brittleness is increased and plastic deformation is suppressed. The phenomenon of significant elastic deflection of substrate is found (pic. 1). Its value may be evaluated from geometrical data (situation of "marks" and imprint) [3]. Substrate is restored elastically till former level after unloading.

MG is destroyed brittle, without plastic deformation at $T_{an} \approx 888$ K. Characteristic zones of destruction arose at $T_{an} \approx 888$ K (pic. 2). The scheme of deformation is shown at pic. 3. In such cases the elastic deflection may reach 50 μm .

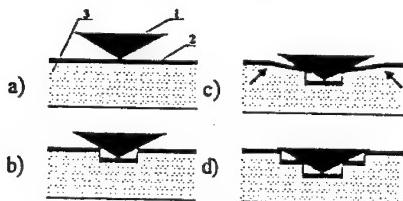




Pic. 1. Elastically restored surface of annealed MG. Arrows show the imprints of Vickers pyramid ribs.



Pic. 2. Brittle destruction of MG annealed at $T_{an}=888$ K.



Pic. 3. Scheme of destruction zone formation during indentation of MG at $T_{an} \approx 888$ K.

Conclusions.

1. The character of deformation and destruction of MG depends on: mechanical properties of MG and substrate; adhesion between MG and substrate, value of load during indentation.

2. The depth of indentor immersion may be established post factum not always because indentation of system MG – substrate is accompanied by elastic deformation of substrate.

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Acknowledgements

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INFLUENCE OF SURFACE ELECTRIC CURRENTS ON THE ALKALI-HALIDE CRYSTALS CONDITIONS

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The structural changes of alkali-halide crystal surfaces were experimentally found. The crystals simultaneously acted by a heat and electric field. The changes were presented as «drops» of a viscous liquid. Is detected, that the occurrence of «drops» is caused by accumulation of an excess electric charge on a surface and local overheating, owing to a bombardment of the surface by ions of air [1, 2]. However, behavior of a crystal, when the electric field is oriented parallelly to a surface was not explored. The purpose posed: to explore influence of complex action of the heating both electric field on the condition and structure of surfaces {001} of alkali-halide, the electric field is oriented of a parallelly explored surface.

In experiments used the single crystals of LiF and NaCl with contents of impurity from 10^{-2} up to 10^{-5} wt.%. The sizes of samples are $20 \times 10 \times 5$ mm. The experiments were carried out in the temperature range from 293 to 893 K. The scheme of the experimental installation is depicted on the fig. 1. Between electrodes the sample positioned. The constant voltage 400 V was enclosed to electrodes.

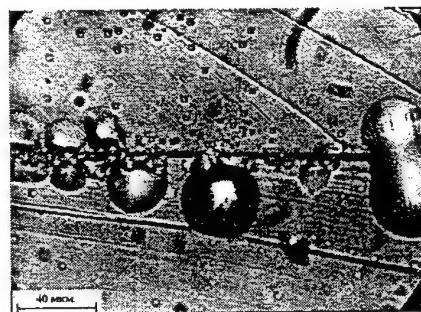
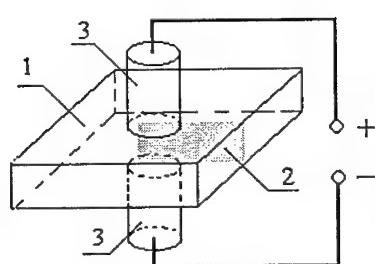


Fig. 1. The scheme of the experimental installation. 1 – the sample; 2 – crack; 3 – electrodes.

In the interval of intrinsic conduction (is higher 823 K) the crack healing was observed. At partial healing, on surfaces of crack, restricting not healed sites, the changes were observed. The changes were exhibited also as drops of viscous liquid (fig. 2.). The researches of a cleavage, which is perpendicular to the crack plane, have shown also presence of changes of interior areas of a crystal near to the crack about 100 microns.



It is possible to explain a healing of a crack, injected into a crystal, during processing of a sample by a diffusion of a material from interior areas of a crystal in a vacuity of a crack. The directional drift of a material is stipulated by temperature difference of a superficial layer of a crack and long-range areas of a crystal. The formation of drops of viscous liquid on the surface of a crack is explained to that on the surface the current intensity is much more, than on volume basis of sample. It can result in to local overheat of superficial areas.

Thus, irrespective of a direction of lines of the electric intensity, the changes of surfaces have similar character and are explained by similar mechanisms.

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ABOUT ORIGIN OF MICROCRACKS ON BOUNDARIES AND IN THE TIPS OF TWINS

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Among of the offered mechanisms of microcracks origin [1] the greatest interest evoke mechanisms based on formation of plane's piling up of dislocations [2] at their stopping by defects of a crystalline lattice. The origin of a microcrack was observed in such case in experiments [3] and take place, as a rule, at merging head dislocations of piling up. During the deforming of metals with BCC and FCC by lattices the important role connected with twinning, which interrelation with fracture was considered repeatedly, however till now remains ambiguous. In [4, 5] was shown, that the account of features of a geometrical structure of twin's boundaries allows to receive close to a real relation between quantity of critical stress of crack's origin and number of dislocations in piling up.

The purpose of investigation – to estimate conditions origin of microcracks on boundaries and in tips of the twins in metals with BCC and FCC by lattices.

We are assumed, that the stopping of the head twin's dislocation calls formation of a stepped piling up of twinnings dislocations in its boundaries. Each dislocation in boundary of the twin displaced on distance equal interplanar (a fig. 1). The stopped tip of the twin was simulated by a double stepped piling up of twinnings dislocations located in points with coordinates $(x_i; y_i)$ (fig. 2). The dislocations with the numbers $i-1$ and i place in the next atomic planes, which are slip planes. Thus $y_i - y_{i-1} = h$, where h is the interplanar spacing. The dislocations in boundaries place pairwise, i.e. the coordinates x_i for $i-x$ of dislocations of the upper and lower boundaries of the twin are equal, and the coordinates y_i are equal on an absolute value, but are opposite on a sign. The head dislocation "is immobile" in a point with coordinates $x=y=0$, and remaining dislocations of piling up are pressed to it by an external stress τ .

On a procedure stated in [6], we determined dependence of energy of a pair bend on an external stress, on which evaluated critical stress and values, according to them, of critical distances between head dislocations.

We have received an analytical expression of origin microcracks in stepped pile-up of dislocations simulative boundary of the twin and the twin, as $\tau = an^b$, where n is the number of dislocations in an piling up, a and b – parameters, τ_{cr} – critical stress. Is shown, that at thermoactivated approach the merging of dislocations in the twin happens more easy, than in isolated boundary.



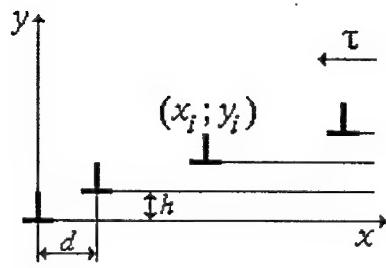


Fig. 1. The plan of dislocation model of the stopped boundary of the twin.

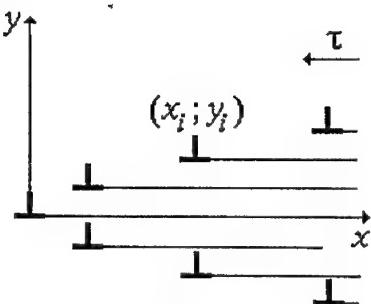


Fig. 2. The double stepped piling up of dislocation simulative the twin.

The analysis of critical's distances values between head dislocations, determined on force and thermoactivated to mechanisms, has shown, that for all considered metals is characteristic thermoactivated the mechanism of origin fracture.

With rise of value of a shear modulus for all considered metals is marked the coming together of critical distances between head dislocations, determined on both mechanisms. We have made the guess, that the character of fracture's origin can vary at other relations of geometrical parameters and elastic constants of materials.

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ELECTROMAGNETIC STIMULATION OF CRACK HEALING IN TRANSPARENT DIELECTRICS

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Strength characteristics of materials are determined by presence of defects: pores, microcracks and so on. However in practice not only detection of presence or absence of crack in material must be done. It is necessary to eliminate the defect, for example to prevent its development by healing.

The purpose of this paper was to investigate experimentally the influence of electromagnetic radiation on processes of stress relaxation and healing of crack tip in alkali-halide.

At the first series of experiment crystals were heated in stove in the temperature range of 300 \div 773 K.

At the second series crystals were illuminated by tungsten lamps. The power of lamps was 20 and 100 W and the maximum energy in the spectrum was 1.06 and 1.24 eV correspondingly. The wavelengths were 350 \div 760 nm. The crystals were heated by illumination till 325 \div 355 K. The temperature of samples was controlled by thermocouple. The intensity of lighting was changed from 4 lx to 15 klx accordingly to power of lamp and lightfilter. The time of illumination was varied from 10 to 1500 hours.

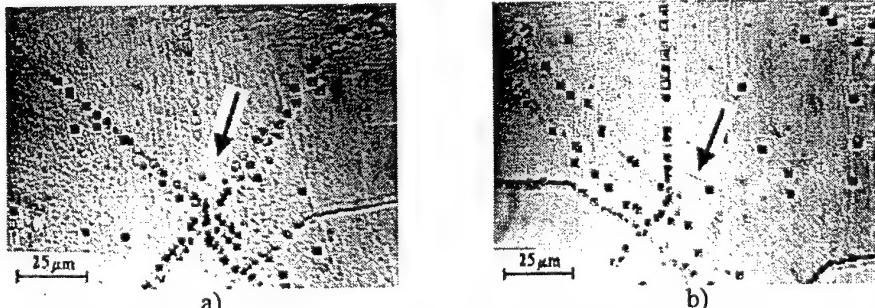


Fig. 1. Dislocation structure in neighbourhood of tip of cleavage crack in LiF single crystal to influence of electromagnetic radiation visual spectrum E=1,95 eV, T=313 K, 15 lx: a) "Control" crystal, T=300 K. b) "Testing" crystal after t=6 hours.

In the third series the crystals subjected to action of radiation with a wavelength $\lambda = 0,154$ nm. An exposure of crystals by X-rays was made by X-ray diffractometer (DRON-2). The time of action varied from 3 about 60 minutes. Site guessed of healing of a flaw subjected to an irradiation by a bundle of breadth ~ 1 mm.



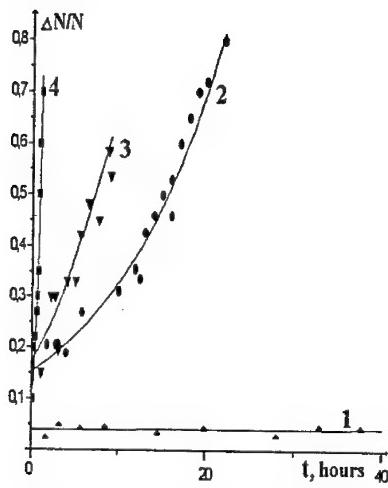


Fig. 2. The relationship between comparative changing of dislocations number at crack tip ($\Delta N/N$) and time (t) of treatment: 1. $T=355$ K, 2. $\lambda=760$ nm, 3. $\lambda=350$ nm, 4. $\lambda=0,154$ nm.

It was experimentally found, that electromagnetic radiation changed dislocation structure at tip of the crack (fig. 1). The summary density of dislocations was lowered. The healing at the crack tip was observed [1, 2].

Electromagnetic radiation causes decreasing of mechanical stresses in tip of the crack of a reversible motion of dislocations and partial healing of the tip.

The exponential dependence of dislocation density both from temperature and from time of lighting was determined (fig. 2).

The intensity of healing and relaxation of stresses depends on material and spectrum of electromagnetic radiation. The processes of healing and stress relaxation depend on intensity of electromagnetic radiation. The greatest effect is observed at action of a X-rays.

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DESTRUCTION OF CAST IRON AT THERMOCYCLING LOADINGS

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Due to low cost, good thermophysical properties and satisfactory mechanical characteristics the cast iron is the basic engineering material for many parts working at thermocycling loadings as well as for parts undergoing rising temperatures in combination with static and dynamic loadings. The ingot moulds for ferrous and non-ferrous alloys, foundry metal moulds and glassforming tools are such typical parts.

An attempt was made to evaluate the affect of the shape of graphite inclusions to resistance of cast iron to fracture under the action of tensile loading at room temperature and under thermocycling loading. The consecutive change of the of graphite inclusions from the lamellar to the globular one was obtained as a result of inoculation of a melt cast iron with increasing additive of magnesium alloying composition.

Both at room and at high temperatures the initiation and propagation of microcracks in a gray cast iron with a flake graphite occurred exclusively on graphite inclusions without a perceptible deformation of a metallic matrix. In a cast iron with vermicular graphite cracks were formed mainly through inclusions normally arrangements to the action of tensile loading. With increasing the extent of spheroidizing of graphite inclusions their participation in the initiation and propagation of cracks was declining, and a globular graphite played the least role in crack formation and cast iron fracture.

A change of graphite shape from the flake to nodular was resulted in change of graphite inclusions form parameter λ_g from 20,3 to 1,4, in a reduce of stress intensity factor for inclusions α_g in 3,5 times, in increase of tensile strength σ_t and bend strength σ_b in 5 and 3 times respectively, in increase relative elongation δ in 50 times. Stress intensity factor K_{Ic} increased in 7 times, heat resistance N increased in 4 times. Experimental data shown to the availability of successive relationship: shape and quantity of inclusions – graphite inclusion form parameter – fracture mechanism cast iron properties.

The possibility to improve cast iron resistance to fracture at cyclic loading enable to enhance the reliability and useful life of ingot moulds, metallic moulds, glass forming tools and different articles made of cast iron and operating in conditions of thermal loadings. Data listed in the table point to the availability of inverse relationship between the graphite farm parameter and the service life of ingot moulds: with a decrease of latter i.e. with the change-over from the gray cast iron to nodular one the useful life of ingot moulds increased in 2.5-3.3 times.



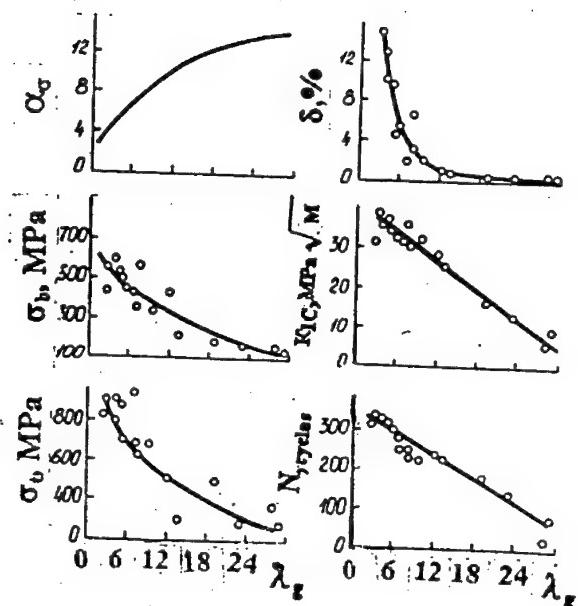


Fig. Relation of cast iron properties to graphite form parameter λ_g

The influence of graphite inclusion form and parameter on the useful life of ingot moulds

Metal poured into a mould	Form of graphite inclusions and λ_g	Service life of ingot mould (number of filling)
Copper alloys	lamellar, $\lambda_g = 18,0$	150
	globules $\lambda_g = 1,42$	500
Aluminum alloys	lamellar, $\lambda_g = 20,3$	2800
	vermicular, $\lambda_g = 8,4$	7200



NORMALLY AND ANOMALY DEFORMATION BEHAVIOUR OF
INTERMETALLICS

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There was performed a comparative analysis and was proposed a description of the different kinds of temperature dependence of the yield stress and work hardening rate in the various intermetallic. The typical feature, that is inherent for the some intermettallics, is the temperature anomaly of the yield stress $\sigma_y(T)$, i.e. his increase with the rise in temperature in the definite interval. The anomaly of the $\sigma_y(T)$, which, at first, was detected in the Ni₃Al alloy, was found later in many alloys with the superstructure of the Li₂ and also with others superstructures. It was found others features of the deformation behaviour of the intermetallics in the field of the anomalous motion of the $\sigma_y(T)$: anomalous dependence of the work hardening rate $\theta(T)$ from the temperature and the weak dependence of the of flow stress $\sigma_y(T)$ from the strain rate $\dot{\varepsilon}$. It is observed the blocking superdislocations in the area of the anomaly behaviour of the $\sigma_y(T)$, and at temperature higher then the maximum of the $\sigma_y(T)$ the octahedral slip systems replaced with the cubical ones

The different experimental dates, which connected with studding of the deformation behaviour and microstructure alloys with superstructure Li₂, give an account and new views about dislocation transformations, which is caused the anomaly in the temperature dependence of the deformation behaviour, is proposed. It was observed that the main elements of the microstructure in the field of the anomaly dependence of the $\sigma_y(T)$ is long straight screw segments, which is blocked superdislocations transformed into a Kear-Wilsdorf barriers. It is shown that exactly these transformations caused the anomaly (positive) temperature dependence of the $\sigma_y(T)$. The unblocking of the indicated barriers determines the transition to the normal temperature dependence of the yield stress at the enough high temperatures.

An explanation was proposed for a nonmonotonic temperature dependence of the yield stress $\sigma_y(T)$ in TiAl having two extrema where $\sigma_y(T)$ changes its temperature behaviour [1]. The comparison of $\sigma_y(T)$ curves for TiAl and typical curves for other materials (BCC metals, semiconductors, Ni₃Al-type intermetallics) allowed reconstructing the shape of the potential relief for dislocations in TiAl. The shape of such Peierls relief reflects existence of two types of dislocation traps (shallow and deep ones) and two types of potential barriers: low and high barriers for the capture of dislocations. The reason of the creation of such relief is the



anisotropy of the charge density distribution near the Ti atoms. The deformation behaviour of TiAl was described over the whole temperature interval allowing for the capture of dislocations in traps and their release therefrom. Expressions determining extrema of $\sigma_y(T)$ were derived. Possible dependencies of the work hardening rate $\theta(T)$ in the region of the anomalous trend of $\sigma_y(T)$ were analysed. Conditions of the anomalous behaviour of $\theta(T)$ were ascertained.

It was proposed that the shape of the potential relief of a dislocation changed in the effective range of a microcrack. The capture of dislocations in deep traps, which is stimulated by concentration of stresses near a microcrack, and the inhibited release of dislocations from the traps up to relatively high temperatures are viewed as a possible cause of TiAl brittleness.

There was analysed the unusual behaviour of the intermetallics after the some prestraining. It was proposed a model, that describes the appearance of the new effects in this case, particularly, the presence and absence of the stress "macrojump" after prestraining. It is taking into account the possibility of the blocking some slip systems and switching on such new systems on the low-temperature stage of the deformation.

It is developed, also, a new approach [2,3] to the description of the plastic deformation, in which the dislocation ensemble is considered as a some population, which evolution is determined as a dislocations multiplication, so as their transformation, and such processes are taking place against a background of the elastic stresses, which are created by this dislocation ensemble. In the scope of such approach it can be described the general character of the stress-strain curves for the different materials. It is explained, particularly, the nonmonotic motion of such curves (yield drop), which can be taking place in the field of the transition from the elastic stress to the plastic.

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DEFORMATION OF ELECTRICALLY DEPOSITED FCC – METALS WITH DISCLINATION DEFECTS

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Electrically deposited *Ag*, *Cu* and *Ni* studied in electronic microscope shows the evidences of disclination defects in their structure [1-2]. The most spread of them are broken boundaries and their dipoles, bands of disorientation.

The broken boundaries of the growth origin have a wide spectrum of disorientation angles θ and can have a dislocation nature ($\theta < 10^\circ$) as well as a disclination one ($\theta > 10^\circ$). The disorientation bands with angles $\theta > 10^\circ$, formed during deposition process (Pic. 1) have with no doubts a disclination nature:

extinction contours at the boundary drop and follow one after one; boundary planes are not always crystallographic. The analysis shows those boundaries and their dipoles actually ends in the crystal. For example, when the reflection angle of the film in a microscope being altered, the extinction contour moves and becomes smooth. That means the boundary ends inside crystal and it has a disclination nature.

As a rule, the deformation process of *Cu* and *Ni* has three stages (II, III, IV).

But if the initial structure has broken boundaries and their dipole configurations, the disorientation bands, which ends inside crystal, then the deformation curve has an addendum stage with low and permanent strengthening factor.

Pic.1 The boundaries that ends inside crystal and their dipoles.

The presence of disclination defects runs to sufficient change of the acoustic emission. Instead of the typical smooth curve in the beginning of the III-d stage of strengthening process the stochastic impulses series of lower intensity and sufficiently greater dispersion of impulse income are observed, that can be seen in the pikes of acoustic charts [3].

The large drops in the AE and high correlation of emission at the third stage are provided by avalanche-like growth of mobile dislocations as a result of the unstable boundaries dissociation. The growth of the mobile dislocations density inside the grain with the presence of disclinations dipoles (which are in the great extent more stable as being compared to broken boundaries) can trigger their motion, that occurs with some time delay. It explains the periodicity in emission, which can be revealed by auto-correlation function (ACF). The least stable dipoles



moves first, the most stable moves later as they require the higher density of dislocations and this explains the increment in the ACF period.

The motion of disclinations dipoles provides observed deformations, and is followed by powerful acoustic emission impulses, causes the fragmentation of the crystal and rough traces on surface.

As soon as dipoles motion are fed by dislocations appeared as the result of growth boundaries dissociation, and as soon as their quantity are restricted, it leads to change in deformation mechanism, the forth stage ends, AE activity comes to zero and the V-th stage begins.

Quite probably the deformation would run in oscillation mode if the constant fast increment of the dislocation density were provided.

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**THE LUMINESCENT POLYSTYRENE COMPOSITIONS MATERIAL
DESTRUCTION AT LONG-TERM OPERATION CONDITION**

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Now polystyrene based compositions are widely used in manufacturing of scintillators, solar concentrators, luminescent materials of different application. The similar materials consist of a matrix (polystyrene) and luminescent components. One of the main requirements to their service properties is the long-time strength. For pure polystyrene based ware this value is studied explicitly enough [1], but much less attention was given luminescent polystyrene compositions material stability to aging and destruction at natural conditions and heightened temperatures. In this connection the analysis of composition samples (size 4x20x200 mm³) material destruction at temperatures 60-80 °C within 3000 hours was performed. The samples breaking point σ_d and transparency L (luminescence light attenuation length) change was investigated (Fig. 1).

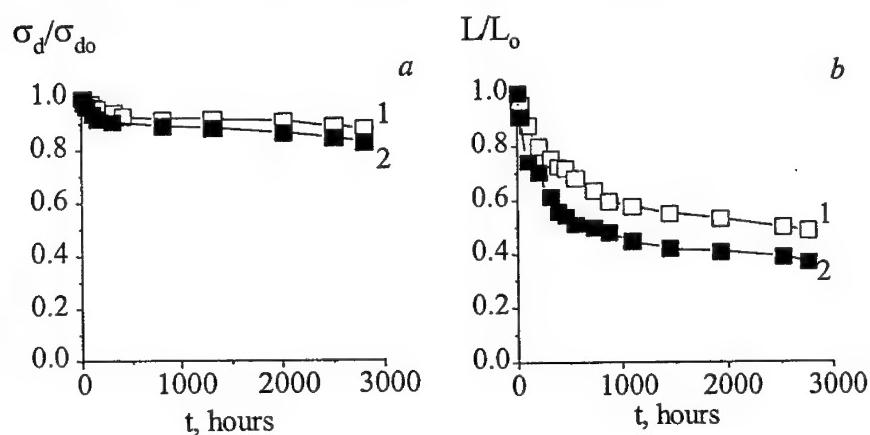


Fig. 1. Polystyrene composition samples properties change as a result of a thermo-aging at temperatures: 1) 60 °C, 2) 80 °C: a) breaking point σ_d ; b) transparency L.

Thermo-aging results in compositions material defects formation, samples painting in yellowish color and microcracks generation (on a surface h=0.1-1 mm, C_c=5-10



cm⁻² and in bulk h=0.01-0.1 mm, C_v=10 cm³), that calls breaking point σ_d change up to 10 %. Besides the defects generated are scattering and absorption centers for photons of luminescence, as Fig.1 show, results in transparency L considerable change of – from above 50 %.

On the basis of performed researches it was established, that at creation of luminescent polymer compositions major factors influential on material frame determining service properties of ware are as follows: matrix molecular-mass distribution selection, plasticizer introduction, different ways of compositions formation usage (polymerization, processing). The breaking point σ_d change (10 %) and attenuation length change (50 %) predictable time is equal 6-12 years depending on methods of their production. Taking into account and usage of the indicated factors allows effectively change both the functional characteristics of a material (strength, transparency [2]), and property of ware, that is stability to destruction in conditions of long-term operation.

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PEARLITE FRACTURE IN DRAWING AND IRON UPSETTING IN CONDITIONS OF PRELIMINARY HYDROGEN EMRITTLEMENT

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Complexities of hydrogen embrittlement processes being caused by collective action of many factors are evident. The phase composition, structure, sizes, morphology of phases, defective substructure and the state of solid solution of material play here the defining role.

It was analized by methods of optic and electronic microscopy. The wire from lowhydrogen steel was undegone to acid removel of scale, then it was calibrated with the range of reduction 0,39 and upsetted with relative deformation 0,67.

Steel drawing is accompanied by formation of line structure, i.e. by streatching of particles of carbide phase in lines. The average length of lines are in two times more than the average longitudinal sizes of ferrite grains; distance between lines are two-three transversal sizes of ferrite graines. Two mechanisms of lines formation in steel drawing are possible. One of them is connected with fracture of pearlite colonies and pilferring of cementite particles by slide dislocations. The second can be connected with presence carbide phase interlayers located on boundaries of grains in steel in initial state. Steel drawing changing the forms of grains brings to placing then in lines.

It is stated that hydrogenation not changing the amplitude level of curvature – fursion structure brings to some in ~ 1,3 times) increase of density of curved extinction contours. This fact testifies about the interaction of hydrogen with distortions of crystalline lattice of matrix. Therefore, in the process of drawing there is the formation of inhomogeneous solid solution of hydrogen in steel.

Drawing of hydrogenated steel is accompanied by formation of micropores and microcracks placed mainly in interphase and intergrain boundaries. The source of their formation is molecular hydrogen and methane gas, being formed as a result of interaction of hydrogen with oxide being extracted in decaying of cementite.

Micropores are placed one by one, the distance between them along the boundary of division is 0,1-0,5 mm. As a rule, the micropores are not the sources of far-acting fields of stresses.

Microcracks generally are acute and in a contrast with micropores they are active. They are the sources of largest curvature – torsion of crystalline lattice. The presence of curved extinction contours at the head of microcracks denote to their deformational origin. It is necessary to note that in hydrogenated specimens the micropores and microcracks after drawing are not found.

Electronic – microscopic research being formed in steel of dislocation substructure showed that the change of deformating type (drawing → upsetting) is accompanied



by increase of volume portion of fragment substructure, by decrease of medium sizes of fragments and by decrease of their anisotropy coefficient, by growing of amplitude of curvature – torsion in crystalline lattice and size of scalar density of dislocations. One can note that the points of bend on presented dependences correspond to change of one type of deformation into the other, namely: drawing – cold upsetting. Intensive deformation, especially in its final stage, has brought to structure fracture of plate pearlite everywhere and to formation of mechanical mixture of α – phase and crushed particles of cementite of irregular form.

Additional factor accompanying the deformation of prehydrogenated specimen is the evolution of micropores and microcracks. First of all there is the increase of density of micropores that brings, in some cases, to their coalescence. The density of microcracks increased greatly and the character of their interaction with material changed, i.e. the crack are relaxed. They are not the sources of far-acting fields of stresses yet. Apparently, the hydrogen being collected during plastic deformation, the carriers, of which are the dislocations, brought to their dulling. The density increase of micropores and microcracks testifies about further leaving of hydrogen atoms and carbon for boundaries. On the one hand brigs to the formation of new weak areas concentrated along innerphase and interphase boundaries of material. On the other hand, it promotes to activating additional mechanism of accelerated fracture of pearlite steel structure of hydrogenated specimens. Namely, leaving of carbon atoms for places of hydrogen collection is accompanied by strengthening the rate of defectiveness of cementite plates, i.e. by formation of weak places in them, on which the cracks during steel deforming can be spread.



INFLUENCE OF MICROALLOING ON CHARACTERISTIC STRENGTH AND PLASTICITIES OF A LOW-ALLOYED CHROMIUM

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Influence of little additives (0,5 wt.%) of zirconium and rhenium on structure and properties of sheet material low-alloyed chromium was studied to rise of strength and decrease inclination of alloys to dynamic deformation aging at the same time increasing technological plasticity.

Successes of studying of dynamic deformation aging and recrystallization embrittlement of chromium and its alloys allowed us to create the new low-alloyed alloys VKh-2K-2, VKhM, VKhMR with raised low-temperature plasticity. Complex alloying of La, Ta, Zr, Re provided high mechanical properties to alloys of this types due to combination of deep refining of matrix and its to solid solution hardening.

Research of structure of VKhM alloy showed that additional alloying by low amount of zirconium (0,1%) led to decreasing of quantity of unconned carbon and to formation of zirconium carbide dispersed thus improving properties of the alloy. A relative elongation at the temperature 1073 K achieved 25-30%.

Distrengthen of sheets of the VKh-2K-2 alloy begins to display after heat treatment at the temperature 1273 K during one hour. However sharp distrengthen of sheets it is observed only after annealing at the temperature 1373 K. A yield of strength is lowered from 680-750 MP \ddot{a} to 280-300 MP \ddot{a} . Necessary to mark that the beginning of distrengthen of sheets does not coincide with the beginning of recrystallization.

It is established that under structure research of annealing samples at the temperature 1223 K, typical structure for deformed state -fibrous structure and fragmentation of polycristalls by cells is remained, the boundaries of cells become more thin. Research of annealing samples showed that recrystallization of the alloy takes place after annealing at the temperature 1373 K during one hour. A grain dimension preserves almost without changes up to 1573 K. Sheet material has the most plasticity after annealing at the temperature 1373 K. Micro hardness herewith falls down to 1500 MP \ddot{a} . Prolonged heat (1373 K, during 10 hours) allows to raise low temperature plasticity of the alloy in recrystallization state. Herewith a relative elongation reaches up to 30-35% due to disintegration processes flowing with formation of disporsion inclusions.



An optimum combination of strength and low-temperature plasticity of sheet material is reached by the following heat treatment: polygonization annealing + recrystallization. Finish heat treatment provides deep disintegration of solid solution and even apportionment of particles of admixture phases of high active alloying elements in metal structure. Duration of heat treatment of ready sheets is determined by the speed of disintegration process and grain dimension and is equals 5-10 hours.



**THE INFLUENCE OF LOADING PARAMETERS AND CONDITIONS ON
LOW-TEMPERATURE INTERMITTENT FLOW AND FRACTURE OF
METALS**

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Low-temperature intermittent flow, which occurs in the elastoplastic region at a temperature of $T \leq 20$ K as a jump-like growth of localized deformation, constitutes an apparent danger for cryogenic structures. The regularities of the manifestation and evolution of this effect depend on many factors among which the loading parameters and conditions are the main ones. The study of their influence was performed individually for each stage of deformation: at the stage of uniform plastic deformation, which can precede the jump-like deformation; in the process of the development of a jump, and directly after its completion. At the first stage, which is subcritical and is characterized by intensive hardening, the system "specimen - loading device" is in an equilibrium state and the rigidity of the system does not influence the deformation behavior of the material. Its influence becomes appreciable when the critical stress has been reached and localized plastic flow begins. The transformation of the system elastic energy into the plastic deformation work and the kinetic energy of moving elements takes place. The lower the system rigidity and the material hardening modulus, the higher is the magnitude of deformation. Experiments at a temperature of 4.2 K revealed that for austenitic steels the twofold reduction in the loading system rigidity leads, on the average to a twofold decrease in the deformation jumps and to a one a half-fold decrease in the specimen relative elongation. When the system rigidity drops below the critical value, fracture occurs in the process of the deformation jumping. Theoretical and experimental studies have been performed into the influence of the strain rate on the process of intermittent flow. An expression has been obtained for the critical strain rate the excess of which does not lead to the occurrence of jump-like deformation. This rate depends on the flow stress, thermophysical characteristics of the environment, and the specimen size. Its calculated values for austenitic steels and aluminum alloys have been confirmed by the experimental results. The authors have formulated the conditions of failure in the process of a jump taking into account the strain rate.

They have studied the influence of the loading conditions, including creep, on the jump-like deformation. In passing from the strain-controlled loading to the stress-controlled one, the number of jumps decreases drastically. As in the case with supercritical strain rate, the magnitudes of the material ultimate strength and relative elongation decrease appreciably. It has been shown that the considerable



influence of the loading parameters and conditions on fracture and mechanical characteristics of metals calls for the modification of the existing codes and standards for testing material samples.



DEVELOPMENT OF STRUCTURAL MATERIALS PROPERTY CONTROL METHOD ON THE BASIS OF RESONANT EFFECTS

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In the solution of a problem of increase of a reliability of products in maintenance the significant role is assigned to development of new process engineering ensuring essential change of properties of structural materials at their constant composition.

As a technological method of control of properties the authors of the present activity offer to use the combined operation of power fields of a various physical nature by reason of that the new structural condition of a material on more high level of ordering [1,2] can be exhibited. Thus they start not from a position of intensity of the operational power sources but from a position of conformity of their frequency characteristics to frequencies of own oscillations of microstructural elements of materials adduced in a non-equilibrium condition. Direct transformation of external energy in resonant activity of own oscillations of microstructural elements of treated materials actually happens.

The object of research selected alloys, widely used in engineering: deformed aluminium-magnesium alloy AMg6M and thermal strengthened carbon steel of general application St-5.

The researches on optimization of modes of external power loading were conducted in 3 stages providing high-frequency processing of alloys in a non-equilibrium condition. The latter was reached at the expense of use of mechanical, temperature and magnetic fields.

At the first stage the influence of electromagnetic oscillations of broad frequency band (from 0,2 kHz to 1,5 MHz) was estimated. At the second stage the operation of impact ultrasonic processing with frequency 20 kHz was studied. And, at last, the third stage provided combination effect on alloys of electromagnetic and ultrasonic oscillations, including impact ultrasonic processing.

The efficiency of external power processing was estimated on change of acoustic activity of alloys, yield point, strength, activity of destruction and time before destruction at tests for long time strength in a mode of a creep. Modes of



accelerated tests of long-time strength were previously perfected. They permit to do in short terms reliable conclusions about change of resource parameters: for St-5 - continuous aging at temperature $T = 400^{\circ}\text{C}$ under a load appropriate to stress 360 MPa; for an alloy AMg6M - continuous aging at temperature $T = 160^{\circ}\text{C}$ under stress 210 MPa.

The optimum mode of combined high-frequency processing of an alloy AMg6M in a non-equilibrium condition raising mean time up to destruction in 3,7 times at tests for long time strength is established. In alloy St.5 the increase of activity of destruction in 1,9 times (short-term.) and in 1,2 times was fixed at electromagnetic loading in a non-equilibrium condition.

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**SIMULATION BY THE STRUCTURE-ANALITICAL MODEL OF
ULTRASONIC VIBRATIONS EFFECT ON STEEL St37
STRESS-STRAIN DIAGRAMS**

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Stress-strain diagrams have been calculated simulating experiments on tensioning of steel St37 specimens when the longitudinal ultrasonic vibrations are simultaneously applied. A model of plastic deformation of polycrystal described in [1] has been used. This model assumes that the material consists of grains, in which the slip systems specific to the given alloy become active when the shear stress reaches a critical value τ^s . The plastic flow goes on under the stress, the increase of which is determined by the hardening modulus h :

$$\tau^s = \tau_0^s + \tau_{def}^s,$$

$$\dot{\tau}_{def}^s = h\dot{\Gamma},$$

where τ_0^s is the initial yield stress level, $\dot{\Gamma}$ is the intensity of the shear strain rate tensor on the slip plane.

Dislocation slip in fcc steel occurs on the 4 crystallographic planes of type {111}. The critical shear stress and hardening modulus were chosen according to the stress plateau and steady-state deformation slope on the experimental stress-strain diagram (fig. 1, curve 1). Intergranular stresses arising during the plastic deformation were also taken into account with the formula $\rho^{gr} = C_p \frac{v}{1-2v} \frac{E}{1+v} (\bar{\varepsilon}_{pl} - \varepsilon_{pl}^{gr})$, where E is the elastic module and v is the Poisson's ratio, $\bar{\varepsilon}_{pl}$ is the average plastic strain of the model object, ε_{pl}^{gr} is the plastic strain of the grain under consideration, C_p is a constant.

Material constants were chosen as follows: $\tau_0^s = 200$ MPa, $h = 250$ MPa, $C_p = 0.0015$, $E = 190$ GPa, $v = 0.33$, density $\rho = 7800$ kg/m³.

The ultrasound influence was considered as an application of a variable tensile stress. The amplitude of stress was estimated as $\sigma_a = E\varepsilon_a = \frac{\pi E \cdot \Delta l_{max}}{\lambda} \approx 13$ MPa, where ε_a is the amplitude of strain, $\Delta l_{max} = 5$ μm is the amplitude of ultrasonic displacement, $\lambda = c/f$ is the wave length,



$f = 20$ kHz is the frequency of ultrasonic vibrations, $c = \sqrt{\rho/E}$ is the sound velocity.

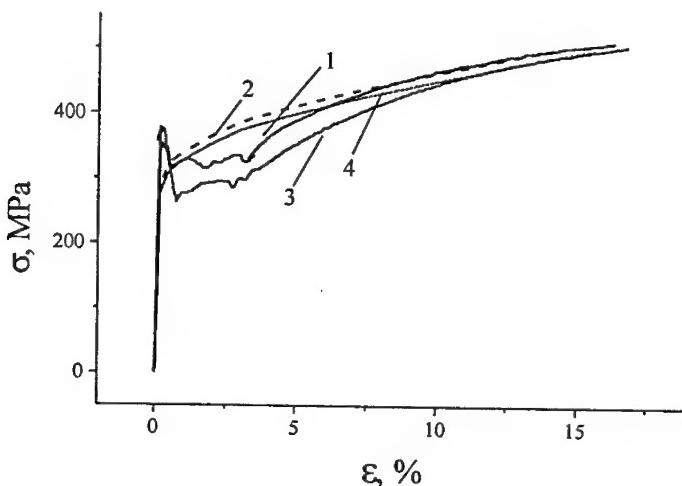


Fig. 1. Experimental (1, 3) and calculated (2, 4) strain-stress curves of steel St37 without (1, 2) and in presence of longitude ultrasonic vibrations (3, 4).

Results of modelling are presented in fig. 1. One can see that the account of ultrasonic vibrations as a variable tensile stress describes the decrease of the flow stress during the deformation.

Acknowledgments

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**NEW CRITERIA OF FLUIDITY OF METALS
UNDER HIGH-SPEED LOADINGS
IN A BROAD INTERVAL OF TEMPERATURES**

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On experimental data [1] about shift fluidity of soft steel under various constant temperatures $195 - 713^{\circ}$ K and velocities of deformation $10^{-3} - 10^5$ s⁻¹, by calculation and construction generalized curve of stress $\tau_p(\xi^T)$, strain $\gamma_p(\xi^T)$, its velocity $\dot{\gamma}_p(\xi^T)$, work $A_p(\xi^T)$ and impulse $I_p(\xi^T)$ of fluidity are analysed, under the concept about "complicated" [2] temperature-temporary (velocity) conformity of its emersion [3], at the temperature of comparison $T_* = 293^{\circ}$ K.

$\lg \xi^T = \lg t + \lg g_t^T$, $\xi_p^T(t_p) = g_t^T(t_p, T)t_p$. Here t , ξ^T - laboratory and transformed (generalized) on temperature time (with an index p - at the moment of fluidity), and $g_t^T(t_p, T) = 1 / g_{\dot{\gamma}}^T(\dot{\gamma}_p, T)$ scale of transformation ($g_{\dot{\gamma}}^T$ - on $\dot{\gamma}$) of time.

$\lg g_{\dot{\gamma}}^T(\dot{\gamma}_p, T) = b(T_* - T) + H[\lg(\dot{\gamma}_p / \dot{\gamma}_* - 3)] \lg \dot{\gamma}_p + c(T_* - T)^d / b$; $b = 0,1875$, $c = 0,125$, $d = -0,33$, $H(x)$ - single function of Hevisaid, equal to zero under $x < 0$. Expression $g_t^T(t_p, T)$ coincides with $g_{\dot{\gamma}}^T(\dot{\gamma}_p, T)$ under replacement $\dot{\gamma}$ for t and $c = 0,101$. In case $T = T_*$, $g_t^T(T_*, t) = g_{\dot{\gamma}}^T(T_*, \dot{\gamma}) = 1$, $\xi_p^T(t_p) = t_p$. The time t_p was determined by an elastic-plastic model: $t_p = \gamma / \dot{\gamma} = (\tau_p / \mu + 0,002) / \dot{\gamma}$, where the module of shift μ was accepted as [4]: $\mu(T) = \mu(T_*)[1 - 447 \cdot 10^{-6}(T - T_*) - 12 \cdot 10^{-8}(T - T_*)^2]$, $\mu(T_*) = 8 \cdot 10^4$ MPa. Work and impulse - $A_p = (\tau_p / 2\mu - 0,002)\tau_p$ and $I_p = A_p / \dot{\gamma}$. The most simple, linear in logarithmic scales, dependences have generalised functions $\dot{\gamma}$ and I_p : $\lg \dot{\gamma}_p = a - \lg \xi_p^T$ ($t_p = 10^a / \dot{\gamma}_p$), $a = -2,4$; $\lg I_p = f - \lg \xi_p^T$, $f = -0,4$ ($-12 \leq \lg \xi_p^T \leq 5$ under $\lg \xi_p^T = \lg(\dot{\gamma}_p / \dot{\gamma}_*)$). The static branches of the remaining functions, under $-12 \leq \lg(\dot{\gamma}_p / \dot{\gamma}_*) < -3$ are also linear in the scale $\lg \xi_p^T$. Their dynamic branches, under $-3 \leq \lg(\dot{\gamma}_p / \dot{\gamma}_*) < 5$, have a much more nonlinear shape with significant growth, especially τ_p and A_p .



For the description of the approach of fluidity under variable load and temperature, the integrated criteria such as "accumulation" of fluidity in the form of damage are offered [5], in which, the generalized curve of "durability" of fortitude to fluidity $\dot{\gamma}_p(\xi_p^T)$ is used as the main supporting function of nuclei:

$$\omega_p(t) = \bar{\omega}_p^{T,t} t = \int_0^t \frac{d\zeta}{\xi_p^T [\dot{\gamma}(\zeta)]} \leq I, \text{ or}$$

$$\omega_p(t) = \bar{\omega}_p^{T,\dot{\gamma}} \dot{\gamma} = \int_0^t \frac{d\dot{\gamma}(\zeta)}{\dot{\gamma}_p[\xi_p^T - \zeta]} = \int_0^t \frac{\ddot{\gamma}(\zeta) d\zeta}{\dot{\gamma}_p[\xi_p^T - \zeta]} = \int_0^t \frac{\ddot{\gamma}(\rho) d\rho}{\dot{\gamma}_p[\xi_p^T - \zeta]} \leq I,$$

Where $\bar{\omega}_p^t$ and $\bar{\omega}_p^{\dot{\gamma}}$ - appropriate operators of accumulation of fluidity. The fluidity occurs, when damage $\omega_p(t) = 1$.

$$\xi^T(t) - \zeta = \int_t G^T[T(\eta), \eta] d\eta, G^T[T(\eta), \eta] = \frac{\partial g_i^T[T(\eta), \eta]}{\partial \eta} + g_i^T[T(\eta), \eta].$$

Equations of an impulse kind have a similar shape, if $\dot{\gamma}$ is replaced for I . The data criteria of fluidity allow to evaluate the medium condition under any monotonous and nonmonotonous, static and dynamic modification of loads and temperatures in a broad interval. The connection of scales g_i^T and $g_i^{\dot{\gamma}}$ allows to install precisely the reverse correspondence of influences of temperature and velocity on fluidity. Transformation allows to realize a significant extrapolation (forecasting).

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CALCULATED DEFINITION OF THE CYCLIC LIFETIME OF THE CONSTRUCTION ELEMENTS WITH FATIGUE CRACK ARRESTED BY COMPRESSIVE RESIDUAL STRESSES.

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The lifetime of the construction element with fatigue crack is known to be considerably increased after the artificial inducing of compressive residual stresses in front of the crack tips. Thus the kinetics of a fatigue fracture allows applying the approaches of a linear mechanics for calculated definition of cyclic lifetime of an element with an arrested crack.

The artificially induced residual compressive stresses σ_{ry} promote the constrained development of a plastic deformation and reduce its dimension. We propose to evaluate the plastic deformation by the following relationship:

$$r_{po} = \frac{1}{2\pi} \left[\frac{\Delta K_{eff}(l_o)}{2\sigma_r + \sigma_{ry}} \right]^2 \quad (1)$$

with $\Delta K_{eff}(l_o)$ being the effective range of the stress intensity factor (SIF) of a cycle and corresponding to the length of the fatigue crack $2l_o$, which is evaluated according to the paper [1].

Accepting the correction of Irvine on a toughness [2] and using the known approach of G.I. Barenblatt [3], the effective range $\Delta K_{eff}^c(l_o)$, which accounts for residual compressive stresses operating in front of the crack tips, can be obtained as

$$\Delta K_{eff}^c = \sigma_{max} \sqrt{\pi(l_o + r_{po})} - 4 \sqrt{\frac{r_{po}}{\pi}} (2\sigma_r + \sigma_{ry}) \quad (2)$$

Then the calculated values of lifetime under the rise of an arrested crack from starting $2l_o$ value up to an intermediate $2l_k$ is evaluated by the relationship

$$N_k = N_o + \sum_{i=1}^k \frac{\Delta l_i}{c_{-1} e^{1.5\lambda} [\Delta K_{eff}^c(l_o + \sum_{j=0}^{i-1} \Delta l_j)]^{m_i}} \quad (3)$$

Here c_{-1} , m_i , and λ are constants of a material, which are defined at tests on cyclic crack resistance of reference specimens.

The proposed method of a calculated estimation of cyclic lifetime has been tried on experimental data for cyclic crack resistance with the specimen made of Steel3 ($\sigma_{0.2} = 296$ MPa, $\sigma_s = 476$ MPa) after arresting of the fatigue crack by artificial inducing of residual compressive stresses by means of dot heat. A flat specimen of large-scale with preliminary grown fatigue crack of 28 mm length was treated by dot heat. The field of residual stresses was formed in a specimen as a result of dot heat. The distribution of biaxial residual stresses along a curve of a crack growth



was examined by a non-destructive acoustic method. The results of the performed measurements are shown in Fig. 1 for one of the specimen ends. Then a sample was tested for cyclic crack resistance at zero-to-tension stress cycle with the maximal stresses of 155 MPa.

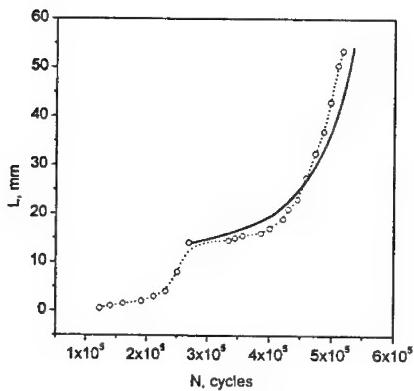


Рис. 2 Dependence of the length of fatigue crack on the amount of loading cycles.

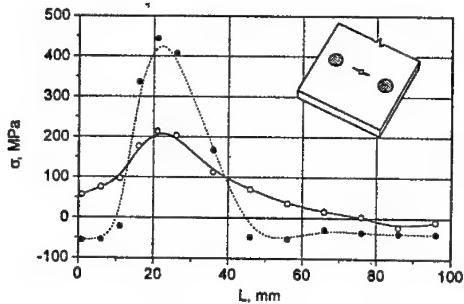


Рис. 1 Distribution of the residual stresses after dot heat treatment
(○- σ_{22} ; ●- σ_{33}).

The lifetime constants necessary for calculation were accepted according to [4] as: $c_1=0.73 \cdot 10^{-13}$, $m_1=3.96$, $\lambda=1.1$. The experimental (dots) and calculated (solid line) dependences of the length of arrested fatigue crack on the amount of loading cycles are shown in Fig. 2. The experimental and calculated results are in satisfactory agreement, as it follows from Fig. 2.

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INFLUENCE OF PHASE TRANSITIONS ON THE SENSITIVITY OF FULLERITE AND POLYMERS TO MAGNETIC FIELD

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Recent experimental studies demonstrate many similar features in physical properties of widely used polymers (polymethylmethacrylate (PMMA), polystyrene (PS), polyvinylchloride etc.) and fullerenes. Molecular type of bonds of these organic solids, mechanical properties of different polymer states of fullerenes [1] and fullerene-containing polymers [2] attract a great attention of specialists. In addition, both diamagnetic polymers and fullerenes are sensitive in their mechanical properties to pulsed magnetic field (MF) [3,4]. The main goal of this work is to study the influence of pulsed MF with amplitude up to 30 T on the plasticity of linear amorphous polymers and C₆₀ single crystals.

General peculiarity of these materials is that they possess great number of phase transitions in the temperature region where the conditions of plastic deformation can be studied more easily. It allowed one to create experimental conditions to step by step «unfreezing» intermolecular relaxations in polymer and to establish that rotational mobility of charged side groups in macromolecule is a necessary condition for pulsed MF influence on PMMA and PS plasticity (fig.1).

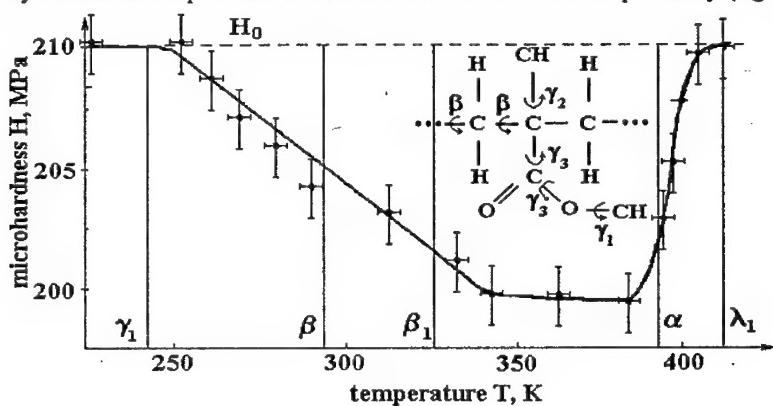


Fig.1. Dependence of PMMA microhardness H on temperature T of exposure of the samples in MF pulse B=25 T. Dashed line shows microhardness of the samples H₀ which were not subjected to the MF action. Greek letters denote temperature of transitions related to «unfreezing» of intermolecular relaxations of the macromolecule shown in the insert.

It was found that the magnetoplastic effect in fullerite changed its sign being close to the temperature of phase transition T_c ≈ 260 K. However, the static

MF does not change plasticity of all materials studied in spite of the fact that magnetoplastic phenomena were previously explained by reorientation of macromolecular bonds which have an anisotropy of magnetic susceptibility [5]. The pulsed MF effect on the plasticity of polymers is equivalent to the combined influence of static magnetic and alternative electric fields with frequency $>10^3$ Hz at T = 293 K.

Thus, temperature and frequency spectra of the magnetoplastic effect were measured. It was shown that these spectra related to inter and intramolecular degrees of freedom determining the sensitivity of polymers and C₆₀ fullerite plastic properties to MF.

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**TENSILE STRENGTH AND VOLUME CONSTANCY
UPON DEFORMATION OF MODEL CRYSTALS CONSTRUCTED
WITH HETEROGENEOUS MUTUALLY PENETRATING BALLS**

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The atomic structure of a solid is presented sometimes as balls, which are connected between themselves by springs. Such a model testifies to capability of solids for resistance to compression and tensile strength, as well. Solids differ from other condensed states by such a capability. No liquid and plasma stretch. Upon compression a solid is known to enlarge and upon tension it becomes elongated towards directions being perpendicular to the direction of an imposed load. The mechanism providing for realization of the Hooke's law and the Poisson's ratio is unclear at present. An energy approach support to describe the known regularities but it does not clarify them.

The model of forming crystal structures with heterogeneous mutually penetrating balls (HMPB) has been suggested in [1, 2]. It is shown that if one draws together HMPB according to the rule of pair mutual penetration, then they are arranged in a fashion that the positions of their centers correspond to the arrangement of the sites of crystal lattices. The lattice type of model crystals formed with HMPB depends upon the relation of sizes of their component parts (an internal ball and a spherical shell). FCC, BCC, simple cubic, tetragonal lattices and diamond or other type lattices may be formed. Lattices of the lowest symmetry category with mutually penetrating heterogeneous ellipsoids are also possible.

The mechanism scheme of HMPB bonding which provides for realization of the Hooke's law and the law of volume constancy under deformation of model crystals is suggested in the given report.

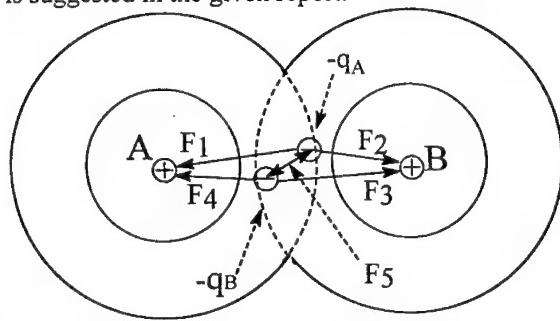


Fig. 1.

Two heterogeneous mutually penetrating balls (HMPB) A и B are shown in Fig.1. The negative charges $-q_A$ and $-q_B$, with equal probability can be at any surface point of the corresponding HMPB. Thus, the negative charge may be considered as equally distributed on the HMPB surface.



Thus, in turn, provides for neutrality to each HMPB. When the negative charges $-q_A$ and $-q_B$ are in penetrated (dotted) sections they simultaneously attract the positive charges of the mutually penetrated HMPB A and B by forces F_1 and F_2 , F_3 and F_4 and, thus, link them. Besides, the negative charges $-q_A$ and $-q_B$, being in the penetrated sections, are repelled one from another with force F_5 and in addition link the HMPB A and B.

If one stretches the HMPB A и B, the distance between $-q_A$ and $-q_B$, when they are in penetrated sections, will reduce. Their mutual repulsion will grow. Tensile strength will also increase. In particular, the tensile strength increase is natural at lengthening for the testing of stretching metals and other materials.

Let us consider a cell, in which the centers of four HMPB are located at the square

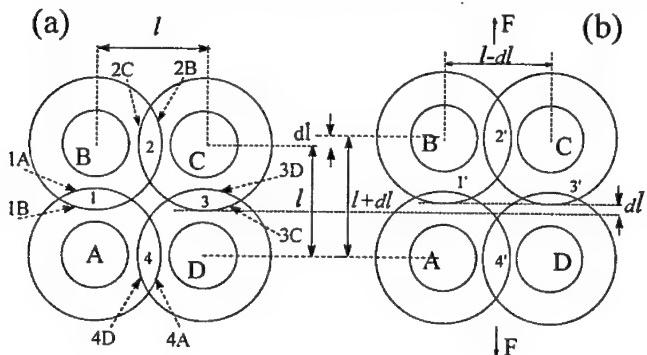


Fig. 2

vertexes, fig. 2a. In the initial state the centers of each pair of the mutually penetrated balls are arranged at distance l from each other and all the penetrated sections 1A, 1B, 2B, 2C, 3C, 3D, 4D, 4A are identical.

As a result of stretching by force F on the value dl the penetrated sections 1A, 1B, 3C and 3D will decrease, fig. 2b. Because of it the repulsive force of the sections 1A, 1B, 3C and 3D from the sections 2C, 4A, 4D, and 2B should decrease accordingly, as well. Then the positive charges of the HMPB A, B, C and D can closer attract to themselves the sections 4D, 2C, 2B and 4A, respectively. It will result in rapprochement of the HMPB in AD and BC pairs, fig. 2b. The cell is narrowed towards the direction, perpendicular to the lengthening, but its volume is saved.

If the cell is compressed the sections 1A, 1B, 3C and 3D will increase and repulse the sections 4D, 2C, 4A and, 2B respectively. The cell will enlarge in the direction of an imposed load. The same regularities are available at deformation of solid-state samples.

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DEFORMATION HARDENING OF COMPOSITE ELECTROCHEMICAL COATINGS

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The composite electrochemical coatings are widely applied in an industry for restoring both hardening of details and tool, and represent a metal matrix with distributed in it by particles of oxides, borides, of nitrides etc. Inclusion of such particles in a galvanic coating allows considerably increasing mechanical characteristics of a coating. On different experimental data the -percent relation of metal of a matrix and ceramic particles in a coating resulting in to the best mechanical characteristics of a material, variously also oscillates from 99,5:0,5 till 67:33. However common legitimacy for all types of coatings one - occurs a limit, at which the further increase of an amount of particles gives in appearance of flaws and losses of strength of a matrix.

The analysis of thin structure of composite electrochemical coatings displays, that at inclusion in a matrix of ceramic particles there are defects of grain formation, namely the value of blocks varies the dislocation density and microstrains of a crystal lattice increases. As a result of derivation around of particles of dislocation closed loops and propagation of crystallizing pressure the yield strength of a material is shifted in the side of increase and corresponds to a yield strength of a material subjected to plastic deformation, when the dislocations, moving under operation of load, will derivate dislocation closed loops.

Apparently, that at increase of an amount of particles in a coating there is also increases of defects of crystalline structure. If in a particular interval of presence of a ceramic phase in a coating these defects render positive influence on connection strength a metal - particle, the further propagation of crystalline defects gives in discontinuity of powers inside a coating, appearance of flaws and losses of strength of a matrix.

The extension of a range of relations the matrix - particle in the side of increase of an amount of particles in a coating, without losses of strength of a matrix, could considerably raise the mechanical characteristics of composite at the expense of greater, on square, contact a detail - hard particle. With the purpose of implementation of this idea it was offered to combine processes of electrochemical moulding of a coating and its plastic deformation. It is known, that the adhesive strength of metal with a particle depends or on presence in metal and particle of like cations or from the value of a plastic deformation on boundary a metal - particle. Usually this value is determined only by value of crystallizing pressure. At usage of a plastic deformation during the process of an electrolysis not only the mechanism of fixing of particles in a coating varies, but also there is an increment of the value of a plastic deformation on boundary a metal - particle, that positively has an effect for strength of composite. Thus the cramping powers in growing



grains of metal increase, that also is favourable has an effect for the mechanical characteristics of composite. Taking root, under operation of an external loading, in a matrix the particle gives in processes originating at plastic deformation of metals, namely driving of dislocations and derivation of dislocation closed loops around of particles. It, alongside with prolonging propagation of a galvanic precipitate, gives in reliable fixing of particles in a coating and possibility of increase of their amount in a composite coating without losses of strength of a matrix. At development of the theory the ceramic particles to an incoherent matrix were taken into consideration.



**A NEW MODE OF DEFORMATION AND REORIENTATION OF
CRYSTAL LATTICE THROUGH LOCAL PHASE TRANSFORMATIONS
IN STRESS FIELDS**

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This paper presents a review of electron-microscopic investigations of the mechanical twins and strain localization bands (SLB's) formed in NiTi alloys, austenitic steels of various types and in a Ni₃Al-base complexly doped intermetallic compound in various conditions of deformation such as cold rolling, torsional deformation in Bridgeman anvils, and deformation in dynamic stress fields generated by high-power ion beams.

In fcc Ni₃Al and austenitic steels, a new SLB type with large-angle boundaries similar to a special Σ_9 -type boundary with a misorientation vector $\theta \approx 60^\circ <110>$ has been detected. These bands are surrounded with regions where the curvature of the crystal lattice is high, which are sources of local stresses approaching the theoretical strength of crystals. Within the boundaries of these bands, traces of α -martensite are detected very seldom.

Analysis of the results obtained has shown that the mentioned bands are formed due to realization of a new deformation mode: bulk compressive-tensile deformation through direct plus reverse (by an alternative system) $\gamma \rightarrow \alpha \rightarrow \gamma$ Bain-type transformations. In this case, the intermediate bcc phase exists only during deformation. The formation of this phase is a consequence of the phase instability of the bcc lattice in high local stress fields and simultaneously serves as a way for relaxation of these stresses leading to a reverse transformation of the bcc lattice to a stable fcc lattice. Alternative schemes of inverse transformation are involved due to the fact that the transformed microvolumes should necessarily change their form through plastic deformation in order that the local stress fields in SLB's to relax.

In austenitic steels, SLB's have been revealed which appeared as a result of multiple ($\gamma \rightarrow \alpha \rightarrow \gamma \rightarrow \alpha \rightarrow \gamma$ type) local structural transformations, including in previously formed strain twins. In the latter case, misorientation boundaries with a characteristic vector $\theta \approx 35^\circ <110>$ are observed, and the plastic deformation can be represented by successive transformations of the $\gamma \rightarrow \epsilon \rightarrow \gamma \rightarrow \alpha \rightarrow \gamma \rightarrow \alpha \rightarrow \gamma$ type. The above values of θ ($\theta \approx (60^\circ \text{ and } 35^\circ) <110>$) are characteristic of the overwhelming majority of SLB's with large-angle boundaries. It follows that for austenitic steels and conditions of mechanical action studied in this work the above mechanism for $\gamma \rightarrow \alpha \rightarrow \gamma$ bulk deformation is the most important mechanism of strain localization.



Analysis of the results of electron-microscopic investigations of strain localization zones in NiTi alloys has shown, that in these alloys the mechanism of direct plus reverse (by an alternative system) $B2 \rightarrow B19(B19') \rightarrow B2$ transformation is the one of the possible mechanisms of (332) and (113) mechanical twins formation.

The above deformation mechanisms provides support for the possibility that deformation may be realized through local structural transformations giving rise to structural states which are several superposed structures or through new permitted states inherent in the electron energy spectrum of the crystal appearing in the internode space. In experiment, these are the nodes of the bcc lattice in the internode space of the fcc phases of Ni₃Al or austenitic steels and nodes of the B19 (B19') martensite lattice in the internode space of the B2 phase of NiTi alloys. It is the motion of atoms through these nodes that is responsible for the bulk compressive-tensile deformation leading simultaneously to the reorientation of the crystal lattice.

It is supposed that the strain carriers in this case are high-energy bulky structures – microscopic regions of a phase unstable in stress fields - rather than linear or plane defects such as dislocations, disclinations, or martensitic phase boundaries.

The $\gamma \rightarrow \alpha \rightarrow \gamma$ deformation bands feature high defectiveness and the formation of structural states such as substructures with a high curvature of the lattice or fragmented nanostructural states with small-angle discrete misorientations. It is supposed that these features are related to the fact that the phase instability in fields of intense local stresses gives rise not only to realization of a new deformation mechanism, but also, because of the decrease in moduluses near the phase transition points, to softening of the SLB material with activation of practically all presently known deformation modes such as dislocation slipping resulting from the decrease in critical shear stresses, diffusion mechanisms for deformation as a consequence of the decrease in energies of point defect formation and migration, and collective rotational modes of mesolevel deformation developing with participation of the above mechanisms.



DEFECT SUBSTRUCTURES AND MECHANISMS OF THE COLLECTIVE MODES OF DEFORMATION IN METAL MATERIALS

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This paper presents a review of electron-microscopic investigations of the mechanisms for plastic flow in metallic materials of various types (with different types of crystal lattice, packing defect energies, inclination to structural phase transitions, original strength, etc.) in various conditions of collective modes deformation activation, such as active tensile deformation of high-strength dispersion-hardened alloys, intense plastic deformation by torsion on Bridgeman anvils and by equichannel angular pressing, cold deformation by rolling, at the front of shock waves generated by high-power ion beams, etc.

The following problems are discussed:

1. Characteristic types of defect substructures of mesolevel deformation which are formed in various conditions of mechanical action, including in zones of stress mesoconcentrators, and their structural models. The most important of them are

- the structural state with *high values of the components* of the lattice curvature rotor or *a high continual disclination density*;
- plane clusters of these defects or *Somiliana dislocations* localized within the misorientation boundaries.

2. Features of the structural states and fields of local internal stresses in zones of stress mesoconcentrators. The parameters of the continual density of defects (dislocations and disclinations) in these zones. High-defect dislocation-disclination substructures as a source of not only intense (approaching the theoretical strength of crystals) local stresses, but also high (up to $G/3 \mu\text{m}^{-1}$) stress gradients (moments), the necessary condition for realization of the rotation deformation mode.

3. New mechanisms for plastic deformation and re-orientation of the crystal lattice:

- *the mechanism of nucleation and motion of partial disclinations* in zones of critical (maximum) crystal lattice curvature and stress mesoconcentrators, which is realized through the formation of substructures with a high continual disclination density followed by their relaxation into Somiliana dislocations or plane clusters of partial disclinations;
- *the mechanism of direct plus reverse (by an alternative system) $\gamma \rightarrow \alpha \rightarrow \gamma$ transformation through bulk Bain deformation*, which is developed in conditions of phase instability of the crystal in fields of intense local stresses and can be described adequately based on the notions about plastic deformation



as a sequence of "nonequilibrium local structural transformations" being modes of relaxation of stress concentrators;

- *flows of nonequilibrium point defects generated during plastic deformation* in fields of high local gradients of stress concentrators.

4. The main physical factors of the activation of new mechanisms for mesolevel deformation, shear-rotation instability, and localization of plastic flow and the appearance of new high-energy strain carrier in stress mesoconcentrator zones most important of which are intense local stresses and high stress gradients and effective suppression of their relaxation by dislocation mechanisms of deformation.



**KINETICS AND MICROMECHANISMS OF MASS TRANSFER ON
INITIAL STAGE OF INDENTATION FORMING AT A LOCAL HIGH-
SPEED PULSE LOADING**

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Recently for examination of bulk properties of solids in submicrovolumes and thin near surface layers the increasing popularity is got by nanoindentation methods. It is caused in many respects by considerable betterment of experimental procedures and hardware basis, which allows to carry out examinations with the high spatial resolution on depth of indenter penetration (up to 0,1 - 0,01 nm). Thus the developed procedures of examination of processes of indentation forming and deformable band about it allow to determine a lot practically means of the performances: quantity of hardness, microhardness or nanohardness of a material; a yield strength; elastic modulus; allocation of impurities and internal stresses; a porosity, thickness both mechanical characteristics of thin films and coatings etc.

Offered by us earlier the procedure of a dynamic indentation, when under activity of the pulse load the kinetics of indenter penetration is explored with the high spatial and temporal resolution, has allowed to explore a kinetics of indentation forming, to reveal division stages at its shaping, to establish their kinetic and thermoactivated of the performance and to spot dominant micromechanisms of mass transfer at material under indenter on each of the revealed stages.

However both in this procedure, and in procedures offered by other authors, the examination of indentation forming begin after a prior indenter contact with a sample surface. It establishes badly checked prior strain of a material in the indentation area.

The initial indentation occurrence and impossibility further to carry out a deforming with major velocity on an initial section of checked loading do not allow to simulate situations with a transient local loading incipient in major number of practically important cases: at an fretting, dry friction, wear and so on. In this connection by the purpose of the present operation was to explore a kinetics and micromechanisms of the incipient state of indentation forming at a high-speed pulsing local loading.

The offered procedure consists in the conferring indenter before touch with a surface of an explored material of some velocity and recording values of effective forces and kinetics of indenter penetration with high temporal resolution (order of 0,3 μ s). The experimental apparatus consist from vertical located rod with indenter, which was hanged on given distance from a sample surface with the help of an electrodynamic drive unit. The cutout of a current in the reel of an electrodynamic



drive unit to set in free slope of a rod with indenter for result. Varying distance from indenter and a sample surface it was possible to set velocity of indenter displacement from zero (indenter before penetration is slightly concerns a sample) up to tens millimetres per one second. It provided velocity of relative deformation on the initial stage of indenter penetration up to $10^4\text{-}10^5 \text{ s}^{-1}$. Values of effective forces was measured by piezo-sensor, which located immediately under a sample, and indenter displacement – with capacitive gauge.

The examinations were carried out on ionic single crystals (KCl, NaCl, LiF) and other materials at ambient temperature. Kinetics of an incipient state of indenter penetration, velocities of relative deformation, value of effective forces, contact voltages and dynamic hardness, and as values of quantity of activation volume is spotted for all explored materials. The features of a local high-speed deforming in requirements of change of relative strain rate on six orders of magnitude (from 0,1 up to 10^5 s^{-1}) are investigated.



**REALIZATION OF UNSTABLE PLASTIC DEFORMATION DURING
THE DYNAMIC INDENTER PENETRATION AT Al-3.3%Mg AND
 $Pd_{40}Cu_{30}Ni_{10}P_{20}$**

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It's a known fact, that a number of materials, including Al-based alloys, such as Al-3.3%Mg, Al-3%Cu, can undergo two kinds of transformation depending on the temperature and speed of macroscopic deformation. One of them is homogenous and continuous deformation, the other - unstable discrete plastic flow. In the second case the curve, that describes relation between stress and deformation, shows a significant number of cases where the local speed of deformation significantly differs from the average speed. The transformation from continuous to discrete flow usually takes place at certain ratio between temperature and deformation speed.

Similar phenomena take place at the deformation of some amorphous alloys, in spite of the differences in their plastic deformation mechanisms in comparison to metals. Their most important common feature lies in the presence of critical speed of plastic deformation, surpass of which suppresses discrete flow of plastic deformation and its temperature dependence.

The present paper deals with investigation of properties and modes of alloy Al-3%Mg and volumetric amorphous glass $Pd_{40}Cu_{30}Ni_{10}P_{20}$ plastic behavior in micro- and nano- representation by means of dynamic indentation. The choice of the methodology is caused by the fact that indentation is one of the most simple and commonly used method of mechanical tests, that allows investigation of material's behavior in high-speed local loading.

The tests were undertaken on the experimental apparatus, consisting of horizontally oriented mobile rod, placed on the elastic suspension rope, that provides its step-by-step movement without friction. On one of the ends a standard indenter (Vickers or Berkovich diamond pyramid) is placed, in the central part - the capacitor gauge of displacement, on the other end - the electrodynamic load drive. The executing of the operations is computer-based.

The load was undertaken by the triangular load impulse with the duration between 0.1 and 400 seconds. Amplitude rates of the load magnitude were designed not to allow possible micro- and nanocracks in the volumetric amorphous glass under the indenter, and for $Pd_{40}Cu_{30}Ni_{10}P_{20}$ varied from 4mN to 85 mN, that corresponds to the maximal depth less than 1 μm , and for Al-3%Mg were designed to provide most effective visual control over discrete plastic flow and varied from 0.1N to 3N, that corresponds to the maximal depth in Al-3%Mg up to 15-20 μm .



Test results, obtained by the means of dynamic indentation, provide information on the imprint formation kinetics and allow to define a number of mechanical properties of materials named above. In particular the type of deformation was registered (continuous or discrete plastic flow), the reasons for deformation leaps, their number, amplitude, duration and attempt of first leap formation, numerical value of static and dynamic hardness was measured, the magnitude of Young's effective modulus was determined too.

It has been proved that with the allowing range of the experimental apparatus, the depth of first leap formation and the number of leaps in Al-3%Mg depend on the indenter penetration speed, and their number increasing with the depth. It has also been proved that the presence of unsteady plastic flow leads to the chaotic change in the direction of sample exit from under the indenter, that causes rectangular contour deformation of the imprint. The quality of the deformation significantly depends on the dynamic properties of the experimental apparatus.

For alloy Pd₄₀Cu₅₀Ni₁₀P₂₀ the deformation leaps appear on the critical depth (~50-100 nm), with growth in amplitude depending on the deformation speed. The number of leaps on the diagram better corresponds to the number of specific traces around the indentation, according to the electronic microscope scanning, than to the greatness of magnitude. At lesser deformation speeds they appeared only during the phase of load growth. Besides, if on the direct load branch they always possess only one sign, on the second, especially on greater speeds, - both. The fact proves that the internal stresses, when they exceed the external tensions, relax during the load process.

To summarize the results, the paper proves, that under local loading, including small loads, during one test two kinds of deformation can be observed, both continuous and discrete plastic flow of material. In spite of the significant differences in structural and mechanical properties of the Al-based alloys and amorphous metal glass, the processes of unstable plastic deformation, resulting from dynamic penetration, reveal great similarity.



KINETICS OF RELAXATION PROCESS OF SUBMICRON LOCAL DEFORMATION IN SOLIDS NEAR SURFACIAL LAYERS

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Nowadays micro- and nanoindentation is widely used for determining mechanical properties of a solids, such as hardness, Young's modulus etc. It should also be considered, that high degree of load localization, high speed of microvolume deformation often significantly transform the properties of the material under study compared to its macroscopic analogues. For instance if while indentation of LiF there used the load force impulse ranging from 0 to 250 mH according to the linear rule with the duration range from 0.1 to 10s, then in the second case the size of the imprint would be 1.15 the size of the sample in the first case. It could be concluded that the reason for the greater indentation under such conditions possibly lies either in the ratio change between the input of various mechanisms during the indentation formation, or in the activation of the new ones. In both cases it could be expected, that different dependencies could be discovered in imprint restoration.

It's a known fact, that after the indenter unloading a partial recovery of the imprint could be observed. The depth undergoes the most changes, and the degree of recovery is determined by different factors, including duration time of its formation with the other conditions remaining the same. To confirm the last thesis the following measurement techniques was undertaken.

The indentation formation and the measurement process of the left over recovery were undertaken consistently with activation of triangular pulse with maximal load 250 mH, with linear acceleration ranging from 50 ms to 50 s, and the regressive time remaining constant at 100 ms. In order to keep the mechanic contact with the imprint while its recovery the constant force of 1-2 mH was applied on the indenter. In the course of tests it has been proved that during the first phase (10-15 s) the recovery, for example, for LiF, of 30-100 nm was registered, and the peaks correspond to time regions of the greater load. The curve, depicting depth reduction, could be sufficiently described by means of exponential function with the duration time of relaxation of a few seconds, that significantly depends, as well as a fore-exponential multiplier, on the duration time of the indentation formation. In the described technique the primary recovery stage, which is supposed to develop after the "instant" load removal, is greatly distorted due to the weakening part of the load pulse of 100 ms, preventing the indenter cast away from out of the imprint as a result of elasticity forces activation. In order to reconstruct the mechanism of indentation depth recovery at the instant load removal, the following technique is used.



It's a known fact, that P-h diagram, based on test with the quasi-static load, allows to determine the energy, used for plastic and elastic deformation at the indentation formation, by means of integrating the corresponding parts. The last batch of energy at instant load removal returns to the indenter as kinetic energy. When comparing elastic force, calculated from the P-h diagram, to kinetic energy of the indenter with its numerical value calculated from the obtained cast away speed of the indenter from the imprint, it became possible to state their significant difference. The utilization of an pulse of a specially designed form allowed to calculate the cast away speed of the indenter as a depth function and thus determine the waste of elastic energy depending on the recovery phase. Assuming the viscous property of the forces, that cause the restoration of the imprint, the obtained data allowed to calculate the speed of uninvolved surface recovery (surface that doesn't experience indenter influence) of the indentation, then a shift function, that becomes an exponent. For LiF the relaxation time has been registered as milliseconds, and, like in the previous case, significantly depends on the duration time of indentation formation.

A greater potential for investigating the changes of imprint condition due to various reasons characterizes the technique, based on the series of consecutive test pulses. Its main idea is that it allows to define the properties of energy dispersion mechanism of the imprint as a pulse serial number function. According to the test objectives, the sinus-form pulses of constant amplitude and period are chosen to perform the function of the testing pulses. The energy dispersion is calculated in a regular fashion using the P-h diagram.

The approbation of the technique yielded the following data: the attenuation characteristics of the energy dispersion in different materials could be successfully described (at least the first 10-20 cycles) by the exponential function with its constants depending both on the type of material and on the duration time of indentation formation.

All of the suggested techniques could be successfully used for solutions of certain problems in relaxation process investigation, dealing with imprint recovery under different conditions of their formation, and on a larger scale could be useful for micro- and nanoindentation investigating.



**INVESTIGATION OF SOLIDS NEAR-SURFACE PROPERTIES BY
DYNAMIC NANOINDENTATION METHODS**

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In recent years nanoindentation is widely used for determine of solids near-surface mechanical properties. It is obviously that mechanical properties and material behavior in nanoscale may be strongly differ from similar conventional macroscopic tests data (uniaxial compression/tension, torsion, bend, creep and so on). There are many reasons for it. Load localization leads to high limitation of plastic relaxation and large hardening of material at deformation zone and so the strain may be approach to theoretical limit (~ 0.1 Young's modulus) even in the soft materials. Nanohardness of some materials can exceed the yield stress τ_y under the uniaxial tests in homogeneous stress field on two or more order of magnitude as a result. Similar conditions are realize in contact zone into many significant cases – friction contact, abrasive wear, fretting, tribochemistry, corrosion, electroerosive wear and others. One of the promising method to create a similar situation under controlled conditions may be nanoindentation with high displacement and temporal resolution simultaneously. Dynamic nanoindentation allows to get over from macroscopic methods of testing (friction, wear, fretting i.e.), which average over to time and space for many contact points, to study of the elementary dynamic act at single short-term nanocontact. However in order to realize dynamic nanocontact interaction under well monitoring conditions it is necessary to combine the high spatial resolution ($\sim 0.1 \div 1$ nm), the high temporal resolution ($\sim 1 \div 100$ μ s) and opportunity to vary some load parameters in wide range. Comercial nanotesters has displacement resolution up-to-date 0,1 – 0,05 nm and slow response, which essentially reduce ability of its application for time-dependent material properties investigation.

In nanoindentation laboratory at Tambov State University was elaborated apparatus which realize nanoindentation conditions with high displacement (0,1 nm) and temporal (0,3 μ s) resolution at load rate range from 10^{-2} mN/s to 2×10^6 mN/s. This apparatus and corresponding software tools have allowed to realize several new techniques of nanoindentation and to carry out the complex research of processes and properties of materials under dynamic nanoindentation conditions. There were: 1) creep under constant load P_0 applied and vibrocreep under harmonic load $P_0 \sin \omega t$ added to P_0 ; 2) time-dependent properties of nanohardness under conditions of deep loading rate change (from 10^{-2} mN/s to 10^5 mN/s); 3) time-dependent properties of indentation depth recovery during unloading (elastic ("fast") and visco-elastic ("slow") components were studied separately); 4) behavior of materials under pulse loading with "instantaneous" and



linear increasing of load at the wide range of loading rate. It helped to determine the atomic mechanisms of mass transfer under indenter including initial stage of the penetration. 5) influence of indentation size on the time-displacement characteristics (homogeneous and unhomogeneous, serrated flow) of plastic yielding processes in materials with different structures; 6) high-cycle fatigue behavior of materials under various rate and shape loading, which approach indentation conditions to real nanocontact interaction processes.

The information abilities, service data, spatial and temporal resolution of techniques elaborated allows to relate they to the modern nanotechnology means. Investigation of dynamic nanocontact interaction is a real way to reveal the scaling boundary of the mechanical properties of materials in micro- and nanovolumes and atomic mechanisms of plastic flow. It is impossible to understand nature of the manifold secondary physical phenomena at high local pressure (phase transitions, various types of the emission, electroconductivity changes, mechanochemical reactions and others) without these data.



USE OF STRUCTURAL MODELS FOR INTERPRETATION OF EXPERIMENTAL RESULTS ON STRETCHING AND BENDING OF SPHEROPLASTIC SAMPLES

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Spheroplastic is one of perspective materials in shipbuilding, rocket production. It consists of a polymeric matrix, in which glass, ashen or polymeric microspheres are distributed at random with high degree of concentration. With the view to detail study of physico-mechanical (including strength) characteristics of spheroplastic, a cycle of experimental studies has been conducted. Stretching and clear bending experiments performed of spheroplastic samples have shown that spheroplastic fracture occurs in a brittle way. Strength limits when stretching and bending were determined by formulae

$$\sigma_B^{(1)} = \frac{P_f}{F}, \quad \sigma_B^{(2)} = \frac{M_f}{W} \quad (1)$$

In (1) P_f , M_f — fracturing force and moment,

F , W — area and elastic moment of sample resistance.

The experiment has shown that $\sigma_B^{(2)}$ is 35% above $\sigma_B^{(1)}$.

This fact is explained in the report by way of structure-statistical interpretation of fracturing process.

As shown in [1], current tension s in the stretched sample for Daniels structural model (corresponding to brittle fracture) is determined by formula

$$s = s[1 - F_r(s)]N \quad (2)$$

In (2) — current tension in non-fractured structural elements;

F_r — structural element strength limits distribution function; N — total number of structural elements in the sample section.

Fracturing stress $\sigma_B^{(1)}$ is determined from the condition

$$s[1 - F_r(s)]N \rightarrow \max$$

If we substitute integral expressions of the bending moment in the section for its quadrature approximation (with accuracy up to 1...2%), then the section will be broken into a row of local areas (stripes). One can suggest that structural elements, composing these local areas of the section, constitute a representative sample from all general population of structural elements. Then, for a sample testing on bending, instead of (2), we will get the relation

$$M = \frac{4N}{H^2} s_0 \int_0^{\frac{H}{2}} x^2 [1 - F_r(\frac{2x}{H} s_0)] dx \quad (3)$$



In (3) S_0 – current tension in structured elements located in the extreme fibres of a sample, the other designations are introduced above.

Fracturing stress $\sigma_B^{(2)}$ is determined from the condition that $\sigma_B^{(2)}$ corresponds to the maximum M.

To study what the difference can be between $\sigma_B^{(1)}$ and $\sigma_B^{(2)}$, these parameters were theoretically established in task F_r by power dependencies proposed in [1]. The received results gave the same divergence level between $\sigma_B^{(1)}$, $\sigma_B^{(2)}$, as observed in the experiment.

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FRACTURE CRITERION OF SPHEROPLASTIC STRUCTURES WITH SLIT RECESSES

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Fractured elements are used in a row of technical systems. Under the influence of load of the given level, these elements must disintegrate into fragments of the given configuration. To secure such controlled fracture, a system of slit recesses (concentrators) is created constructively. A method of determining the load fracturing an element is discussed in the report. To determine the load, it is necessary to establish a criterion of cracks appearing in the vertices of recesses. Criterion establishing methodology is given with reference to spheroplastic structures. The following postulates are taken as basics:

- according to the criterion Rice, a crack appears in the vertex of a concentrator, if maximum stretching stresses reach some critical meaning σ_c in the so called process zone with size r_c (r_c - typical microstructural size of material);
- stresses in the area of the concentrator vertex are determined on the assumption of elastic asymptotic of stressed deformed state, proposed in [1].

Taking into consideration these postulates, critical meaning of stress σ_c is determined by relation

$$\sigma_c = \frac{(1 + \frac{\rho}{\rho + r_c}) K_1}{\sqrt{2\pi(\frac{\rho}{2} + r_c)}} \quad (1)$$

In (1) ρ – radius of a recess blunting, K_1 – intensity stress coefficient calculated for fracturing load and matched a perfect crack having the same length as the recess.

Analysis of the experimental data and structural properties of spheroplastic allows to suggest that ρr_c . Then from (1) we get

$$\sigma_c = \frac{2K_1}{\sqrt{\pi\rho}} \quad (2)$$

A series of experiments have been conducted on samples of spheroplastics. Flat long samples with transverse recesses have been subjected to stretching till fracture. With that, length and radius of the recess were varied. Stresses σ_{ci} were determined by results of the experiments with the use of (2), their meanings were processed statistically.

Variation coefficient of a random value σ_{ci} was 12%.



Verification of hypothesis H_0 on equality of average meanings σ_{ci} when testing different types of samples has shown that with a significance level of $q=0.05$, hypothesis H_0 is true. That is, one can consider that the above-mentioned approach to establishing a spheroplastic fracturing criterion in the concentrator zone is being proved experimentally. Relation (2) can also be interpreted in the following way [2]:

$$\sigma_c = \sigma_{y\max} \quad (3)$$

where $\sigma_{y\max}$ – normal stress in the vertex of a concentrator. Using finite elements model and determining the load by way of criterion (3), bringing about developing of a crack along the recess, all process of their fracture has been studied for a row of spheroplastic structures. Calculation has shown good agreement with the experiment.

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SMART MATERIALS APPLIED FOR SENSORS OF DEFORMATION DAMAGE IN COMPLEX ALLOYS

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The deformation damage control and lifetime forecasting in materials which are used in time-dependent conditions are very complex and challenging questions. Under these conditions sensors are expedient that allow us to forecast lifetime of the most loaded components of constructions without their destruction and deformation damage analysis. For this purpose we used Al-single crystal sensors because they are the most perspective for aircraft needs where many components of constructions are exposed to fatigue impact.

The single crystal plates with thickness <200 μm and dimensions ~10×25mm were fixed on complex alloy sample by means of glue PASCOFix (which proposed in Germany) for mechanical testing. The loading of sample with single crystal plate (sensor) was carried out in elastic deformation interval of basic sample only. While the sensor was deformed plastically during this loading, because plastic deformation of sensor is one of the necessary conditions for appearance of sensor surface relief. Analysis of deformation slides (relief) allowed us to obtain the qualitative and quantitative description of deformation damage basic material.

The sensors with orientation <001>{100} were established to be the most suitable for practical use. Fig.1 illustrates the typical surface relief on this sensor and its evolution during loading.

For characterization of very intensive evolution of the sensor surface relief we applied the fractal analysis which is applicable for objects with self-similarity on different scale level, namely, information fractal dimension [1].

It is obviously that for application of fractal geometry one can need to find the scope of self-similarity for property investigated. This task is highly dependent on physical sizes of the system, because mathematical definition of self-similarity in relation to physical objects is highly concerned with the certain size limitations. For example, the lowest scale level for digitized images is theoretically limited by the lowest value optical resolution of scanner device or digitizing resolution. In our case it was equal to one pixel. The highest scale level is limited by the size of digitized image, which was equal to 500–600 pixels for single image and 1500–2500 pixels for panoramic view.

We calculated information fractal dimension for the specimens with unknown location of strain localization site. These specimens were loaded under different conditions (see Fig.2). From these simplest dependencies of fractal dimension vs. the number of cycles for some stress amplitudes we derived conclusions as to the



current deformation pre-history. We proposed several criteria which can be used on practice, for example, the simplest is as follows: the maximal value of fractal dimension can be a sign of minimal lifetime for the given deformation conditions and the material used. Moreover, the maximum drifts to the lower numbers of cycles along with an increase of stress amplitude. In that way we can use the information fractal dimension for quantitative characteristic of deformation relief in sensor surface.

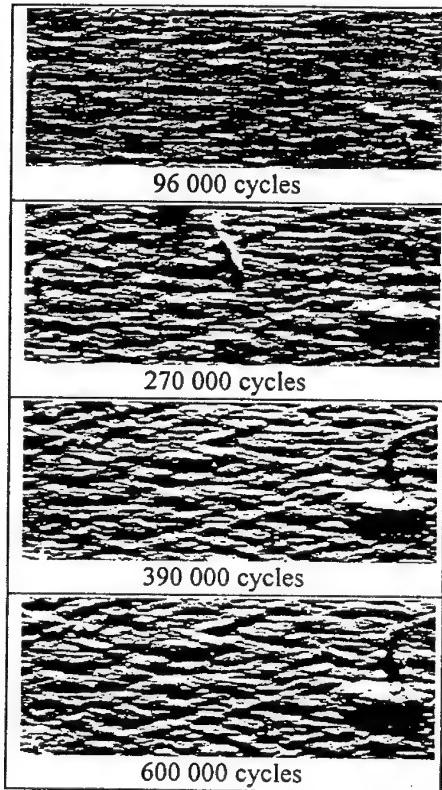


Fig.1. Evolution of sensor surface relief ($<001>\{100\}$, 146 MPa, 512×200 pixels)

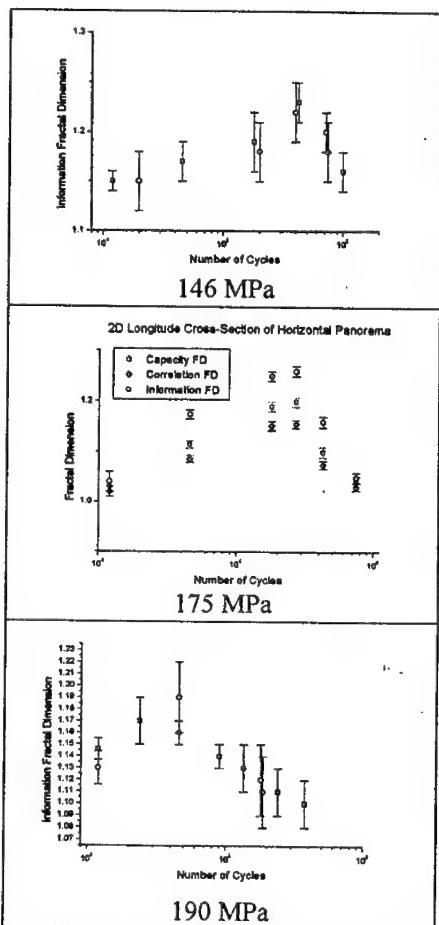


Fig.2. Evolution of fractal dimension

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THE ELASTIC AND INELASTIC PROPERTIES OF SUPERPLASTIC
MATERIAL CHANGING UNDER EXTERNAL MECHANICAL STRESSES
ACTION CONDITIONS

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The problem of superplasticity takes the important place among the hardness and plasticity physics problems. The questions about the mechanism of superplastic deformation of finegrained materials and the nature of the structure state that provides for superplastic (SP) behavior of typical SP alloys with small grain belongs to the actual ones of this problem. So, the investigations of elastic and inelastic properties of SP materials proved to be important because these properties of SP materials remain practically unresearched.

In this paper there are the results of investigation of inelastic aftereffect, internal friction and Young's modulus changes depending on the applied stress magnitude σ and the duration of natural alloy aging for typical SP finegrained alloy Pb-62%Sn during deformation. The molten samples followed by compression with the hydraulic press have been investigated. The mechanical deformation was carried out in the creeping regime and under external cyclic loading. The interval of the applied σ included both σ optimal value and σ differing from it directed to increasing and decreasing. The experiments were carried out at the room temperature when the alloy in one of investigated states shows its superplastic properties.

It was stated, that the inelastic aftereffect deformation depends essentially on the applied stress magnitude. The value of this deformation is maximal when the value of σ is optimal for demonstrating superplasticity.

The character of plastic flow and the inelastic aftereffect differ essentially depending on the alloy aging duration. The plastic flow under creeping condition of recently compressed samples and being aged for nearly three months is characterized by the presence of accelerated creeping phase, which then becomes stationary. The flow velocity and the aftereffect deformation magnitude of new-prepared alloy exceed essentially those of the aged alloy. The inelastic aftereffect deformation magnitude has the same peculiarity. The correlation between the character of plastic flow and inelastic aftereffect is observed during the different phases of creeping. The creeping of the samples aged for some years is performed at continuously decreasing velocity. The magnitude of inelastic aftereffect deformation decreases along with the creeping development.

It was established, that in the initial samples the relative quantity of Sn- and Pb-phases changes from 1:1 in the new-prepared alloy to 1,9:1 after nearly 2.5



years aging. During the deformation the volume ratio of phases in all cases changes approaching to the equilibrium one, which, as the calculations tell, is 2.5:1.

The analysis of the experimental data mentioned above enable us to assume that one of the reasons defining both of plastic and inelastic properties of the alloy investigated is the peculiarities in phase transformation taking place under external σ action, in particular, decomposition of supersaturated solid solution. Both of the maximum of internal friction $Q^{-1}(t)$ discovered inside the 0 - 25 minutes interval of cyclic deforming of the samples aged for a long time and alloy Young's modulus changing during deform fixed in these experiments speak in favour of this assumption.

The results of the X-rayodiffractometric researches, which were carried out by means of ДРОН-3 using filtered CuK α -radiation, reveal the changes of the crystal lattice parameters of alloy phases during deforming. This change may be the evidence of decomposition of supersaturated solid solution based on Pb under external loading conditions. The X-rayogram peculiarities of the alloys investigated at the initial state in the area of Sn(200)-, Sn(101)- and Pb(111)-reflexes point to anomalous diffusive X-rays dissipation. Such peculiarities in the samples deformed are not observed.

The complex of investigations leads to the assumption that superplastic effect in the Pb-62%Sn alloy traditionally referred to the alloys which show structural superplasticity, may be connected with the phase transformation taking place under the deforming conditions.

It should be noticed, that the results of investigations carried out, in particular, X-rayostructural analysis display that an important role in manifestation of high plastic properties of the alloy is played by the internal stresses, originated in the samples due to mechanical compression of ingots.



FRACTURE MECHANISMS OF POLYMETHYLMETHACRYLATE UNDER DYNAMIC LOADING

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There has been a great deal of work invested in the analysis of fracture surface. Much of the interest in this branch of engineering, called fractography, is concerned with the determination of the location of the onset fracture of a given structure together with the probable cause for its failure. Although every fracture surfaces is different, the proven utility of fracture surface analysis in the determination of different fracture processes stems from the fact that a close relation exist between the deterministic dynamics of crack and the fracture surface that it leaves behind. In many cases the mechanisms leading to characteristic surface features are not known. It is one of the goals of the study of fracture to uncover these fundamental mechanisms and to understand their generality.

Crack propagation is investigated experimentally in polymethylmethacrylate (PMMA) under shock loading created impulse magnetic field [1]. Plain specimens of PMMA 10 mm thick had notch 3 mm width and 10 cm length. Speed photo camera SFR-2 was used as photo register. Pressure amplitude constituted $P=140-320$ MPa.

From many studies of fracture surfaces formed in brittle materials, it was found that the surface created by the process of dynamic fracture has a characteristic structure in brittle amorphous materials. This structure, called "mirror, mist, parabolic surface markings, hackle", has been observed to occur in materials as diverse as glass, ceramics, noncrosslinked glassy polymers such as PMMA and others [2].

Using microstructural investigation the dynamic fracture surfaces of brittle amorphous material was analyzed depending on distance from notch tip. Propagating from the notch tip fracture character changes from fiber, quasi-fiber to cup fracture like fracture types in metals. Diagram of different kinds of fracture depending on impulse duration and distance from notch tip was built. This dependence has periodical character under loading of microsecond duration impulses of threshold amplitude.

The fracture energy, or the energy needed to create a unit fracture surfaces, is of tremendous practical and fundamental importance. For this reason, evaluation of the fracture energy as a function of the crack velocity have been done. Surface fracture energy was evaluated, using cup size and specific surface fracture energy. Its dependence on distance from notch tip was defined. Results are presented in Table 1.



Table 1

Crack length, L_{cr} , mm	Cup size, d , μm	Crack velocity, V , m/s	Surface fracture energy, E , J/cup
5,0	32	92,4	$17 \cdot 10^{-8}$
17,5	18	50,696	$5,3 \cdot 10^{-8}$
25,5	13	194,25	$2,8 \cdot 10^{-8}$

The formation and evolution of meso-branches are a major influence of the dynamics of a crack. The existence of mesocracks that branch away from the main crack was discovered only in the places with fiber and quasi-fiber fracture types. They did not find in the places of cup fracture. But character of mesocracks differs from ones presented in [2]. Inspection of the fracture surface in a given experiment shows that mesocrack inclination angle changes nonmonotonously depending on the distance from notch tip. Mesocrack inclination angles and places of their nucleation were determined. As for crack velocity, it has nonmonotonous character depending on distance from notch tip similar to changing of fracture character but differ from data presented in [3]. These results are presented at the Table 2.

Table 2

Crack length, L_{cr} , mm	Crack inclination of mesocrack, °	Crack velocity, V , m/s
1	32	271,6
3	33	119,8
6,5	29	70,58
7,1	19	58,95
7,6	28	62,27
27	26	126,79

Conclusions.

1. Diagram of different kinds of fracture was built.
2. Fracture surface energy was evaluated.
3. Mesocrack inclination angles and places of their nucleation were determined.

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INELASTIC PROPERTIES OF METALS TREATED WITH PULSES OF LOW MAGNETIC FIELD

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The conventional technologies of material processing no longer satisfy increased demands of production process. To realize a resource of properties remained after such processing, the materials and products are influenced by pulses of low magnetic field. However, this problem has been studied insufficiently, in particular this concerns the inelastic properties of metals. This article deals with the above problem.

The inelastic properties such as the low-frequency internal friction, ultimate strength σ_B , yield strength σ_T , plasticity δ were investigated during the micro- and macroplastic deformations of metals prior to, during and after the action of low magnetic field pulses (LMFP) with $H \leq 10^5$ A/m.

The objects of investigation were specimens of wire from different materials: brass, aluminium, zinc, copper, tungsten, molybdenum, steel. Diameter of the specimens was 0,15 to 3 mm and the working length varied from 50 to 100 mm. The LMFP treatment was performed at a standard «ОИМП-101» unit. Low-frequency internal friction was investigated by the reversible torsion pendulum method with the oscillation frequency of about 1 Hz. The mechanical properties were determined by means of tensile tests.

It is shown that during and after the action of low magnetic field pulses the level of the dislocation internal friction in brass, aluminium, steel and zinc is changed. During the action of pulses the changes are instantaneous and those after the magnetic treatment go in two stages: at first they are the short-term ones (1-5 hours) and then they are durable (for 5-12 days and nights). Their kinetics is of different character (monotonous, nonmonotonic, damped oscillation) and finite values tend either to the initial or to the obtained during the magnetic field influences.

The mechanical properties (σ_B , σ_T , δ) of steel, brass, tungsten and molybdenum were investigated before, just after and in 7 days after the magnetic influence. It has been found that LMFP influence practically all of the investigated materials, but the brass. The properties can be either improved or made worse. For the obtaining of a more complete information on the character of changes in the mechanical properties the behavior of the ultimate strength and yield strength of steel-70 samples were investigated for a long time (45 days and nights) after the influence with low magnetic field pulses. The above values were changing in a similar nonmonotonic way in the form of lasting oscillations with increasing amplitude and periodicity.

The observed changes in internal friction are associated with those in the defect-impurity complexes of the impurity-dislocation type, whereas changes in the



mechanical properties are, most probably, related to a change in the character of interaction between moving dislocation and crystal-lattice defects (impurity atoms, aggregates of impurities, crystalline-grain interfaces, stacking faults) induced by pulses of low magnetic field.



**PRELIMINARY STRAIN INFLUENCE ON
DEFORMATION BEHAVIOR CHARACTERISTICS
OF SUPERPLASTIC ALLOYS**

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The model of preliminary strain influence on superplastic behavior of alloys is described. The quantitative model describes the influence of low prestrains, which do not change the grain size (there is no refinement) but change grain boundary state. The model is bases on ideas and models of the theories [1, 2, 3] what allows to describe the strain rate and temperature of preliminary deformation and not only prestrain value. During the superplastic deformation the dislocations carrying out intragranular slip introduced into grain boundaries form a defect layer. This defect layer causes stress fields influencing on intragranular dislocations motion. On the other hand, dislocations introduced into grain boundaries and spread there. These defects change the free volume of the grain boundaries, and their diffusion properties are changed respectively [2].

Thus, the prestrain results in: 1) additional stress internal fields appearance (these fields retard of the lattice dislocation motion inside the grain) and 2) changes in grain boundaries diffusion properties. All that causes changes in characteristic dislocation spreading time and relaxation time of internal stress. The first of the indicated factors - stress fields - is the reason for hardening [4], the second factor - changes of diffusion properties - may be the reason for softening (after prestrain). Each factor may be dominant, depending on temperature and strain rate condition, and also on strain value and prestrain conditions. Competition of these two factors is the reason for complicated character of prestrain influence on superplastic behavior of materials.

Obtained theoretical results are compared with experimental data for two superplastic alloys: Zn-22%Al and Pb-38%Sn.

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**OXIDE AND CARBIDE DISPERSION STRENGTHENED POWDER
COPPER OF "DISCOM" TRADE MARK: DEFORMATION AND BREAK
FEATURES**

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Different types of Oxide and Carbide Dispersion Strengthened Powder Copper (OCDS-Copper) developed by Scientific & Technological Company TECHMA Ltd, possessing unique characteristics set and considerably higher performance resource under extreme conditions in comparison with traditional copper alloys, are applied in numerous technical fields, in particular – as electric contact parts of welding machines (electrodes for resistance spot welding, contact tips, etc.); electric contacts, sliding contacts being among them; guiding sleeves and valve seats, etc. OCDS-Copper types are supplied both as ready-made goods, and as semi-finished goods, e.g. rods, tubes, bars and so on. Different methods of mechanical processing, turning, milling, grinding in particular, and also different methods of metal processing by means of pressure are used when producing parts. However, structure peculiarities of the OCDS-Copper types mentioned stipulates the necessity of a more careful approach when choosing deformation methods and schemes.

OCDS-Copper structure of Trade Mark DISCOM is formed in the process of its production, which consists of mechanical alloying of copper powders, carbon and one or several oxide and carbide forming elements in a ball attritor, obtained granulated stuff cold compacting into briquettes and hot extrusion of the briquettes into semi-finished goods (rods, tubes, bars and so on). The stuff structure formed as a result of such a technological redivision can be classified as one directed fibrous, where the fibers are granules stretched during thermal deformation processing, the structure of which in turn is dispersion strengthened. Fiber sizes and their distribution depend on technological parameters of the stuff manufacturing process. As is known, the stuff fiber structure results in characteristics anisotropy and its own features of deformation and break. In particular the stuffs with the structure mentioned show little resistance to interlayer shift and transverse tear, which also concerns OCDS-Copper types of Trade Mark "DISCOM".

Consequently, it is rather urgent to make the right choice of loading schemes when working with the parts made of OCDS-Copper types mentioned and methods of their production.

For example, cubic electrodes for contact welding of steel framework for ferroconcrete items with their fibers placed in the plane of the steel framework being welded and besides transverse to the contiguous to the electrodes rods from which the framework is being welded possess the resource 1,2 ... 1,8 times higher than the resource of the same electrodes placed when welding with the fibers perpendicularly to the plane of the framework being welded.



Thread on the stiff made by means of cutting when fibers are cut through has a solidity 8...10 times less than the same thread made by means of rolling when the fibers obtain thread profile without being cut.

When performing cold stamping of contact welding electrode caps, say of Ø16 mm it is impossible to use the rod of Ø14 mm as is accustomed when producing electrodes from traditional bronze. When settling a workpiece made of OCDS-Copper Ø14 mm per Ø16 mm micro cracks may appear which result in electrode caps cracking. It is necessary to use the rod of Ø17 mm from which a cone workpiece of Ø17 mm and Ø16 mm is made by means of reduction on the first stage. Then an electrode of Ø16 mm is produced from this workpiece on other stages of the process.

Fiber direction should be also taken into account and corresponding technological regimes should be chosen when OCDS-Copper types are processed mechanically. For instance, when performing superdeep drilling of fine holes in contact tips for MIG/MAG welding the drill can be shifted from the drilling axis if we use ordinary drilling machine, but not drilling equipment specially developed for that aim at a definite speed and drive.

A bushing lower part tear may occur when a guiding sleeve made of OCDS-Copper pressed into the Aluminum cylinder block of the engine is bored in the wrong way when a borer comes out.

When stuff structure is not taken into account at cutting splits may happen when the cutter penetrates the stuff.

Many years experience of OCDS-Copper of Trade Mark "DISCOM" research and development, ready-made goods produced from them and their operation created safe recommendations paying attention to these stuffs deformation and break features.



INFLUENCE PLASTIC DEFORMATION ON LONG-RANGE ORDER IN ALLOYS WITH $L1_2$, $L1_2(M)$, $L1_2(MM)$ SUPERSTRUCTURES

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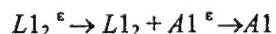
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The order - disorder thermal phase transition has been the subject of inquiry for a large body of experimental and theoretical research. The plastic deformation is capable to give rise to the order-disorder transformation. Essentially this transformation has not been studied although its possibility has been repeatedly mentioned. However, little is known about the effect of deformation on the degree of long-range atomic order, even though indirect evidence on the distraction of long-range atomic order upon deformation has been reported time and again in literature. In recent years information has been obtained that the ordered state vanishes as alloys are ground in ball mills.

The experimental facts concerning the influence of the plastic deformation on the long range order in the well-ordered alloys with $L1_2$ and $L1_2(M)$, $L1_2(MM)$ superstructures have been reported in this work. The alloys Au_3Cu , Ni_3Fe , Ni_3Al , Cu_3Pd , Au_4Zn , nonstoichiometric Cu_3Pt , were chosen for present investigation. The samples were deformed by cold rolling and then studied by the X-rays diffraction methods.

It was obtained the deformation-induced order - disorder phase transformation $L1_2 \rightarrow A1$ takes place and it occurs heterogeneously. The disordered areas emerge after a permanent deformation of a few percents and their amount increases in the process deformation. At rather large strain the alloys become disordered. The lowering of the long-range degree of an alloy is accompanied by an inhomogeneous decreasing the antiphase domains sizes, in parallel with the large domains presence small-sized domains are found out and their proportions grow with the strain increasing. The observed facts give proof to the heterogeneity of the order - disorder transformation, connected with the strain localization.

The effect of the strain on the state of the alloys is carried out under the following scheme:



The heterogeneous behavior of the order - disorder phase transition results from peculiar deformation processes of localized character. Indeed, the formation of a considerable amount of the disordered phase is understood solely in terms of strain-localization concepts. It is believed that high concentrations of strain-



induced defects (dislocations, point defects, antiphase boundaries, stacking faults, etc.) are generated in strain-localization regions characterized by high shears. It is the joint action of these defects, which is capable of diminishing the long-range order parameter down to a critical value below which the long-range order spontaneously reduces to zero.

The disordering mechanisms initiated by plastic deformation are classified into two groups. One group involves the mechanisms that primarily change the long-range order in the gliding planes of dislocations because of the generation of APBs by moving dislocations (or stacking fault). The following processes may cause APB generation: the intersection of moving dislocations with thermal APBs appearing upon ordering; the intersection of superdislocations belonging to various slip systems; the multiplication of superdislocations (this increases the dislocation density and, therefore, the area of APBs between the partial dislocations of the superdislocations); and the dissociation of superdislocations into isolated partials under the effect of local internal stresses.

These mechanisms, which primarily disturb the order strictly in the plane of dislocation motion, may further cause volume changes in long-range order. Taking into consideration wetting of APBs and decreasing of the long-range order on them expression for measured long-range order was obtained. A contribution of the multiplication of superdislocations to lowering of the long-range order was estimated. A change of the long-range order associated with edge dislocations climb was calculated.

Other group of mechanisms is responsible for the long-range order decrease in the volume of deformed crystal. These mechanisms are primarily connected with the point defects generation in the course of plastic deformation.

Plastic deformation, leading to elastic distortions of the crystal lattice, may change the temperature of the order-disorder phase transformation, and the resulting nonuniform stresses may cause decomposition of the solid solution. Factor such as local heating of the material in slip bands also cannot be ruled out.

To clarify the role of these mechanisms in various alloys, models of the deformation-induced destruction of long-range order should be constructed.

It seems likely that the long range order decrease in the ordered materials caused by the joint action of the mentioned above mechanisms.

It may be supposed that in different alloys these mechanisms may manifest themselves differently and may lead to different dependencies of the degree of long-range order on the degree of deformation.



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